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STRENGTH AND PLASTICITY

Cold-Resistant Steels: Structure, Properties, and Technologies

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Abstract—The domestic achievements and research results in the field of the creation of low-carbon, cold-resistant, weldable steels supplied in the form of sheet metal for the shipbuilding industry—in particular, for use in the construction of icebreaker hulls, tankers, gas carriers, ice navigation vessels, and offshore stationary platforms—are reviewed. The requirements of the Russian Maritime Shipping Register for the chemical composition, assortment, and mechanical properties of cold-resistant, structural, low-carbon steels, their certification, and the current state of affairs in the development of cold-resistant steels in Russia and abroad are presented. The features and principles of the alloying and microalloying of low-carbon, cold-resistant, weldable steels of various strengths in the range from 315 to 960 MPa and the mechanisms to harden and enhance resistance to brittle fracture at low temperatures are reviewed. The requirements and technological methods for the formation of a steel structure providing the required cold resistance are formulated, the features of the cold-resistant steel structure are described in detail, and recommendations are formulated for further improvement of the properties and quality of cold-resistant steels.

Keywords: cold-resistant steel, Arc index, phase transformations, thermomechanical treatment, quenching from hot rolling, tempering, mechanical properties, performance, crack resistance, structural parameters, ferrite, bainite, martensite, electron backscattering diffraction pattern (EBSD) analysis

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INTRODUCTION

Cold resistance characterizes the ability of a material to resist brittle fracture at low temperatures. The advent of welded structures (in particular, all-welded hulls of ships and vessels) led to the problem of the prevention of their fragile failures, which were numerous at first. By 1970, more than 250 accidents were recorded on ships with welded hulls; ships often collapsed unexpectedly in calm weather.

In Russia, Davidenkov, Trefilov, Razov, Uzhik, Shevandin, Shulte, Shimelevich, Kroshkin, Gorynin, Meshkov, Gol'dshtein, and other scientists were involved in the prevention and study of fragile steel and welded joints at low temperatures. The reasons for fragile failures, the criteria to assess usability, and methods to assess the propensity of steel to brittle fracture at low temperatures were established.

The main way to prevent the brittle failure of structures is to choose materials that exhibit sufficiently high plasticity and cold resistance at the minimum operating temperature (design temperature), and the main criterion is usually the value of the work (energy) of the impact of small samples at a given test temperature (usually 20–40°C below the operating temperature). The requirements for the latter criterion increase with an increase in the thickness of sheet metal.

The steels 09G2, 09G2S, and 10KhSND were created in Russia in the 1940s. They provided resistance to fragile failure at temperatures down to –20°C and were successfully used for a long time as shipbuilding materials. The first domestic, cold-resistant, shipbuilding steel, SKhL-4 (which is designated in modern documentation as D40S and is similar to steel 10KhSND in its chemical composition) was implemented in the industry in the postwar period and was supplied to the industry after quenching with tempering. It was used to construct large ships. Subsequently, in the 1970s, the construction of nuclear icebreakers, lighter carriers, and shipyards, which became crucial to the problem of year-round navigation along the Northern Sea Route, led to the development and application of stronger cold resistant steels.

In the 1970s and 1980s, the manganese–nickel and chromium–nickel–molybdenum steels 10GNB and AB-1 (12KhN2MD) with a cold-resistant threshold to –60°C [10, 11] were developed at the KM Prometei Central Research Institute (which is now part of the Kurchatov Institute National Research Center). They were produced in limited batches with the use of energy-intensive technologies. They were used in the construction of 12 self-propelled drilling rigs of the Caspii type, ten semi-submersible platforms of the Shel'f type for the Barents Sea and Okhotsk Sea, and in the hulls of the nuclear icebreakers 50 Years of Victory, Taimyr, and Vaigach. The high-strength steel AB2R (12KhN5MDF), which has a limit of no less than 690 MPa for rolled sheets up to 130 mm in thickness, was developed based on these steels. It was intended for use in the rack of self-elevating drilling rigs (which are used in the self-elevating drilling rig Arkticheskaya).

The subsequent research in the 1980s enabled the development of structural and shipbuilding steels that were cold-resistant to operating temperatures of –20°C (in some cases even to –40...–60°C) with a yield stress of 390 MPa or higher. Pridantsev, Malyshevskii, Legostaev, Solntsev, Odesskii, and other scientists dedicated their efforts to the creation of cold-resistant steels [10–18].

Abroad, the production of structural steel for civil shipbuilding in the 1960s and 1970s was developed in the United States, Japan, Canada, Norway, South Korea, Sweden, Finland, and some other countries. The most experience was accumulated in Finland, which built icebreakers and ice-class ships, including those built at the request of the Soviet Union.

Cold-resistant steels of different grades and degrees of alloying, such as the construction and bridge structural steels 09G2S, 10KhSND, 17G1S, 15GF, 16G2AF, etc. [15–18], were used in the early 2000s in domestic industry. In addition, shipbuilding steels with a guaranteed level of the impact work (energy) at a temperature of –40°C were supplied in accordance with GOST 5521–93 (State Standard of the Soviet Union). However, the characteristics of these steels were lower than the foreign requirements for metal quality, the level of physical and mechanical properties, processability, and other parameters. In addition, the scope of controlled properties was also different.

This inconsistency was fully manifested during the design stage and at the beginning of the construction of an offshore, ice-resistant, drilling platform for the Prirazlomnoe oil reserve. The standard technical documentation requirements included in this project substantially exceeded the requirements of GOST 5521–93 (State Standard of the Soviet Union), while Russian industry did not implement the mass production of cold-resistant steels with the guaranteed level of the impact work (energy) at a test temperature of –60°C (category F[1]), which were necessary for the upper structures of the platform.

The cold resistance and crack resistance of shipbuilding steel was mainly increased via additional alloying with nickel in combination with chromium and molybdenum and improvement of the heat-treatment regimes. However, an increase in the level of alloying and the use of energy-intensive technological processes led to a substantial increase in cost, which is especially problematic due to the huge material capacity of structures for the Arctic shelf.

Thus, more than 70 000 t of cold-resistant, easily weldable, heavy-duty steels (with standardized yield points of 315, 355, and 390 MPa) are necessary only for the construction of the supporting part (caisson) of stationary drilling platforms of the Prirazlomnaya type. The used amount of cold-resistant, weldable steel with yield points of no less than 355, 390, and 500 MPa is in the range of 1000–6000 t per icebreaker and up to 5000 t per tanker. Steels with a guaranteed flow stress of 500 MPa or higher are necessary for exploratory, floating, self-elevating, drilling rigs and floating cranes used for the installation of stationary platform sections in the Arctic environment. In general, the demand for cold-resistant steel for Arctic structures, nuclear icebreakers, ice-type vessels, and coastal infrastructure construction is many tens of thousands of tons per year.

An analysis of the operating conditions of the structures on the shelf of the northern seas of Russia was performed by the Krylovskii Scientific Center based on the results of Arctic expeditions. It showed that the following factors typically have a combined effect on the marine oil and gas production and exploration platforms: low temperatures (a design temperature down to –50°C); wind and wave loads (from June to October) with wind speeds up to 47–59 m/s and high waves up to 15–17 m, which cause cyclic loads (up to 10^7 for the operational lifetime); dynamic loads from seismic (on the Sakhalin shelf, up to 8–9 on the Richter scale) and ice-induced (heavy ice conditions from November to March, inclusive, with a flat ice thickness of 1.5–2.0 m) vibrations; repeated static and cyclic loads caused by the operation of the oil-rig platform; and the prolonged corrosive effect of sea water and ice.

The development of offshore oil and gas fields and the creation of structural steels began abroad in the 1950s (in Canada, Norway, the Mexican bay, and on the Angola shelf), but it is impossible to use the international experience fully in the creation of materials for marine structures in Russia, since the climatic conditions on the shelf of Russian seas are significantly more complicated for the operation of structures.

The prevention of brittle failures of structures of marine stationary platforms was of great importance from the very beginning of the development of the northern sea basins, and the formulation of the most severe requirements for the cold resistance of structural steels apparently prevented the negative experience associated with such failures. Factors that objec-

[1] For heavy-duty and high-strength shipbuilding steels, the following notations are applicable depending on the temperature of the samples tested for the impact bending strength: A for 0°C, D for –20°C, E for –40°C, and F for –60°C.

tively increase the risk of brittle destruction of marine stationary platforms in relation to ship constructions include the use of a component with a large thickness (up to 50–70 mm for flat and truss structures and up to 150 mm for racks of self-elevating drilling rigs) and an increase in steel strength. However, the successful operation was achieved only for open sea basins (Northern and Norwegian Seas) with minimum (calculated) operating temperatures no lower than –10°C and for structures built in the Beaufort Sea (T_c = –20°C). The need to develop Arctic basin fields, in which temperatures reach –45 to – 50°C (Pecherskoe Sea), and offshore fields of the Far East region (–35 to –40°C) required a special approach to the creation of cold-resistant steels.

The following high-strength steels were used abroad in icebreakers and marine structures: STE 690 and NAXTRA-70 from ThyssenKrupp; the Japanese steels N-Tuf, HT60, and HT80; steels of the HSLA type (United States); Swedish steels of the OX type; steels of the RAEXE POLAR type (Finland), etc. [19]. The most widespread structural steels are HSLA and A537 (United States), Grade 50 (England), HSB55 (Germany), and Welten 80 (Japan); they have guaranteed flow stresses of 335–500 MPa and cold resistances of –10 to –60°C [20, 21]. In contrast to Russian steels, these are usually manganese steels with a carbon content of no more than 0.18% with small additions of nickel (0.1–0.6%) and copper (0.1–0.3%) that are microalloyed with niobium, vanadium, and titanium simultaneously or in different alloys.

Despite the high reliability of the domestic, cold-resistant, structural steels developed at the beginning of the 21st century, they were not economically competitive with foreign analogs. In fact, this led to the dismissal of domestic metallurgical producers—as occurred in the Sakhalin-1 and Sakhalin-2 projects—and the transfer of design tasks to foreign companies, who used their own technical requirements for applied materials that often did not meet the requirements for reliable operation under the extremely severe climatic conditions of the Russian Arctic shelf.

All further efforts of metallurgical scientists and practicing technologists of metallurgical plants in the period of 2000–2020 were dedicated to the elimination of this inconsistency, in particular, with respect to the development of technologies for the production of a new generation of cold-resistant steels for shipbuilding in Russia. They are considered in this review.

STATE OF THE MATTER

The main requirements for shipbuilding steel are currently standardized by GOST R 52927–2015 (State Standard of Russia) [22] and are based on the requirements of the Russian Maritime Register of Shipping (RMRS), which are coordinated, in turn, with the requirements of the International Association of Classification Societies (IACS). These requirements were basically formulated in the middle of the 1960s and have been essentially rewritten and supplemented regarding the operating conditions of marine equipment under extreme conditions in recent years.

In accordance with the requirements of the IACS, the Rules for the Classification and Construction of Marine Vessels (part XIII, "Materials"), and the Rules for the Classification, Construction, and Furnishing of Marine Stationary Platforms [24] of the RMRS, steels of the standard, enhanced, and high strength categories with a guaranteed flow stress of 235–500 MPa and standardized values of the impact work measured on samples with an acute cut at test temperatures of –20 to –60°C (cold-resistance categories D, E, and F)—which are supplied in accordance with GOST R 52927–2015—are currently used in the building of vessels, icebreakers, and marine technical facilities that operate at low temperatures.

Steels with a higher strength (category 620–960 MPa) are supplied in accordance with the technical standard specifications in agreement with the RMRS.

The standard strength category of steels (with a guaranteed flow stress of 235 MPa) includes manganese steels (≤1.0 wt % Mn) with aluminum, vanadium, titanium, and/or niobium additives for the refinement of grains and can contain small additions of nickel (up to 0.4 wt %). The carbon content in steels is limited to 0.21 wt % for steels of cold-resistance category D (tested at a temperature of –20°C) with normal strength and to 0.18 wt % for E-category steels (tested at a temperature of –40°C). The carbon content in steels that belong to higher strength and cold-resistance categories (enhanced strength category with yield points in the range of 315–390 MPa and the high strength category with yield points of 420 MPa or higher) is limited to ≤0.12 wt % and can be reduced depending on the delivery conditions. At the same time, nickel, copper, and chromium (in small amounts, molybdenum) can be used as alloying elements in these kind of steels.

The content of harmful impurities, such as sulfur and phosphorus, in these kinds of steels is limited to no more than 0.005 and 0.010 wt %, respectively. The most stringent requirements for impurity elements in Russian standard technical documentation are set for cold-resistant steels and steels with improved weldability.

Table 1 gives the requirements for the chemical composition of shipbuilding steel in accordance with the requirements of Rules of the RMRS [23, 24].

The requirements for steels for operation at low temperatures (cold-resistant steels of category F) in the Rules of the RMRS are described in a special section. In 2012, the RMRS included the Arc index (for use in the Arctic region) in the requirements for steels. According to the Rules of the RMRS, Arc is a symbol

Table 1. Requirements [23, 24] for the chemical composition of shipbuilding steel (in wt %)

Strength category	Steel of standard strength	Steel with enhanced strength		High-strength steel	
Cold-resistance category	A B D E	A–E	F, Arc	A–D E–F after TMCP	A–D E–F after QT
C_{max}	0.21 0.21 0.21 0.18	0.18	0.16	0.16 0.14	0.18
Mn_{max}	2.5 × C 0.80 0.70 0.60	0.9–1.6	0.90–1.60	1.0–1.7	1.70
Si_{max}	0.50 0.35 0.35 0.35	0.50	0.50	0.60	0.80
Al_{min}	0.015	0.015	0.015	0.02	0.018
P_{max}	0.035	0.035	0.025	0.025 0.020	0.025 0.020
S_{max}	0.035	0.035	0.025	0.025 0.010	0.015 0.010
Cu_{max}	–	0.35	0.35	0.55	0.50
Cr_{max}	–	0.20	0.20	0.50	1.50
Ni_{max}	–	0.40	0.80	2.00	2.00
Mo_{max}	–	0.08	0.08	0.50	0.70
Nb	–	0.02–0.05	0.02–0.05	0.05	0.06
V	–	0.05–0.10	0.05–0.10	0.12	0.12
Ti_{max}	–	0.02	0.02	0.05	0.05

that is added to the logo of steel brands that undergo a series of tests according to the RMRS program to determine additional cold- and crack-resistance characteristics that meet the requirements imposed on steels with improved weldability. Next to the Arc index is the calculated (minimum) operating temperature T_d[2] of the material (without the minus sign) up to which the steel can be used for any structural elements without restrictions. The principal difference between steel with the Arc index and F-category steel is that the performance characteristics of rolled sheets and welded joints of such steel must be guaranteed by the technology of their production.

Steel can be supplied after normalization (N), controlled rolling (CR), thermomechanical processing (TMP), or quenching (from furnace or rolling heating) with tempering (QT and $Q_{rh}T$, respectively). It should be noted that thermomechanical processing or quenching with tempering is currently used to produce cold-resistant steel. As a rule, normalization does not provide cold resistance at test temperatures below −20°C (for sheets with a thickness of more than 40 mm), and the thickness for controlled rolling is generally limited to a mere 25 mm.

The modern requirements for the magnitude of the impact work of the sheet steel, which are common to all shipbuilding constructions with insignificant deviations to one side or another, can be expressed by the following relations:

$$KV_L\ [\mathrm{J}] \geq 0.1\sigma_{0.2}\ [\mathrm{MPa}], \quad (1)$$

$$KV_T \geq 0.7\, KV_L, \quad (2)$$

[2] T_d is the minimum calculated operating temperature of the material, up to which the steel can be used in any structure components without restrictions.

Table 2. Requirements for mechanical properties of steels

Category	Flow stress R_{eH} or $R_{p0.2}$, min, MPa	Ultimate strength R_m, MPa	Relative elongation A_5, %	Relative thickness contraction Z_z, %,	Impact work, KV_T, J
			no less than		
D, E, F32 (Z25, Z35,W), arc*T*d	315	440–570	22	35	50
D, E, F36 (Z25, Z35,W), arc*T*d	355	490–630	21	35	50
D, E, F40 (Z25, Z35,W), arc*T*d	390	510–660	20	35	50
D, E, F420 (Z25, Z35,W), arc*T*d	420	530–680	19	35	80
D, E, F460 (Z25, Z35,W), arc*T*d	460	570–720	17	35	80
D, E, F500 (Z25, Z35,W), arc*T*d	500	610–770	16	35	80
D, E, F620 (Z25, Z35,W), arc*T*d	620	720–890	15	35	80
D, E, F690 (Z25, Z35,W), arc*T*d	690	770–940	14	35	80
D, E, F750 (Z25, Z35,W), arc*T*d	750	800–970	14	35	80
D, E, F890 (Z25, Z35,W), arc*T*d	890	940–1100	11	35	80
D, E, F960 (Z25, Z35,W), arc*T*d	960	980–1150	10	35	80

where KV_L and KV_T are the impact-work values at test temperatures, which are obtained on samples cut along and across the rolling direction, respectively.

Regarding the establishment of standard requirements for impact work, the Rules of the RMRS are aligned with analogous systems for the selection of materials for the construction of marine equipment of other classification societies and national standards. The main differences between the Rules of the RMRS and other documents of this sort are related to the need to justify the selection of materials for a calculated temperature below –30°C, which is typical of the weather conditions of the Arctic and Far East seas of Russia.

Table 2 gives the requirements for the mechanical properties of cold-resistant steels, including those with the Arc index, in accordance with GOST R 52927–2015 (State Standard of Russia) and technical standard specifications in agreement with the RMS.

The requirements for the value of the impact work exceed the corresponding requirements of Rules of the RMRS and foreign standards.

Additional requirements for steel plasticity are set according to the Z parameters: the relative contraction Z on samples cut in the direction of the sheet thickness should be more than 25 (Z25) or 35% (Z35).

The main factor causing the formation of layered fractures is the presence of nonmetallic inclusions of the MnS type—and SiO_2 and, to a lesser extent, Al_2O_3—in the form of small, deformed petals or strongly deformed plates in the sheets. In addition, certification tests conducted by a metal manufacturer to obtain a certificate of recognition from the RMRS include tests on the propensity of steel to undergo strain aging (which determines the impact work *KV* on longitudinal and transverse samples after deformation and isothermal exposure to a temperature of 250°C).

The temperatures of the ductile-to-brittle transition determined on large test samples, i.e., T_{cb}[3], DWTT[4], and the zero-plasticity temperature ZDT[5], as well as the critical crack tip opening displacement (CTOD) are used as performance characteristics for the base metal (BM) and welded joints.

Tables 3 and 4 give the requirements imposed on steels with the Arc index for the cold-resistance characteristics with respect to the critical temperatures T_{cb} and ZDT, and the crack resistance with respect to the CTOD at a temperature of T_d for the BM and the metal in the heat-affected zone (HAZ), respectively.

The need to use cold-resistant steels of various strength categories is determined by the region of operation and the platform type, which can be a floating platform (MODU, semi-submersible, and self-elevating drilling units), a gravity (stationary) platform supported on the seabed, a drilling platform (limited operating period), or a mining platform (year-round operation).

The hull-structure BM is selected based on the thickness, the design temperature, and the role of the

[3] T_{cb} is the critical temperature of brittleness (ductile-to-brittle transition), at which at least 70% of the fibrous component is observed in the fracture of a full thickness sample with a concentrator in the form of a notch under three-point static bending before fracture.

[4] DWTT is the critical temperature of brittleness (ductile-to-brittle transition), which corresponds to 70% of the fibrous component in the fracture of a full thickness sample with a sharp notch subjected to destruction under impact loading at a speed of 5–8 m/s (test with a falling load).

[5] ZDT is the critical brittleness temperature (zero plasticity temperature) defined as the maximum temperature, at which a standard size sample breaks down with a brittle weld and a notch initiating the crack under impact loading.

Table 3. Requirements for critical temperatures T_{cb} and ZDT for steels with the Arc index [23, 24]

Rolled metal thickness, mm	ZDT, °C					T_{cb}, °C				
	Arc20	Arc30	Arc40	Arc50	Arc60	Arc20	Arc30	Arc40	Arc50	Arc60
25–30	–35	–45	–55	–65	–75	–20	–30	–40	–50	–60
31–40	–40	–50	–60	–70	–80	–5	–15	–25	–35	–45
41–50	–45	–55	–65	–75	–85	+5	–5	–15	–25	–35
51–60	–50	–60	–70	–80	–90	+5	0	–10	–20	–30
More than 60	–50	–60	–70	–80	–90	+5	0	–5	–10	–15

Table 4. Requirements for CTOD values for BM and HAZ for steels with the Arc index [23, 24]

Thickness, mm	CTOD at T_d, mm, no less than									
	BM					HAZ				
	F32W^{Arc} F36W^{Arc}	F40W^{Arc} F420W^{Arc}	F460W^{Arc} F500W^{Arc}	F620W^{Arc}	F690W^{Arc}	F32W^{Arc} F36W^{Arc}	F40W^{Arc} F420W^{Arc}	F460W^{Arc} F500W^{Arc}	F620W^{Arc}	F690W^{Arc}
25–30	0.15	0.15	0.20	0.20	0.25	0.10	0.10	0.10	0.15	0.20
31–35	0.15	0.15	0.20	0.20	0.25	0.10	0.15	0.15	0.20	0.25
36–50	0.20	0.20	0.25	0.25	0.30	0.10	0.15	0.15	0.20	0.25
51–70	0.20	0.25	0.30	0.30	0.35	0.15	0.20	0.20	0.25	0.30
Above 70	0.25	0.25	0.30	0.35	0.35	0.15	0.20	0.20	0.25	0.30

considered unit (the severity of consequences in case of destruction), which are determined by the category of the unit (special, main, or secondary) or the connection group (for drilling rigs). Rolled steel and its welded joints must be certified for compliance with the requirements of the Russian Maritime Register of Shipping, and the supplier must have RMRS approval for steel production, which is regulated by procedures aligned with international requirements.

The results of certification tests of ship-hull steels previously supplied in accordance with GOST 5521–93 (State Standard of the Soviet Union) showed that it is an extremely complicated problem to achieve the set requirements for the crack resistance of welded joints of these kinds of steels.

Since the high tempering of welded structures is impossible due to the large amounts of heated metal in the HAZ during welding, the change in the physical and mechanical characteristics of the steel—especially, the corrosion resistance and mechanical strength—should be minimized.

The main structural factors determining the level of crack resistance of a welded joint from low-alloy steel were determined in [25]. It was established that there is a tendency toward an increase in the $CTOD_{HAZ}$ values with refinement of the initial austenite grain and a decrease in the volume fraction of bainite of lath morphology in the coarse-grained section of the HAZ of the welded joint.

To increase the guaranteed level of crack resistance of the metal in the HAZ of the welded joint, the values of the carbon equivalent C_{eq} (by 20–30%) and the crack-resistance parameter P_{wm} during welding must be substantially reduced for a substantial decrease in the hardness of the metal in it; they are indirect indicators of weldability [26, 27] and are defined as follows:

$$C_{eq} = C + \frac{Mo + Cr + V}{5} + \frac{Cu + Ni}{15} + \frac{Mn}{6}, \%, \quad (3)$$

$$\begin{aligned} P_{wm} &= C + \frac{Si}{30} + \frac{Mn + Cr + Cu}{20} \\ &+ \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B, \%. \end{aligned} \quad (4)$$

Due to a decrease in the quench hardening, a decrease in the contents of carbon, manganese, chromium, and molybdenum, together with the regulated content of microalloying elements, leads to the minimum increase in the hardness in the HAZ of steels with enhanced and high strengths, which reduces the tendency of cold cracking. At the same time, a decrease in the content of elements that form carbides and a reduced carbon content give rise to the homogenization of austenite during rapid heating prior to welding, an increase in the chemical homogeneity in the HAZ of the welded joint, and a uniformity of phase transformations during cooling after welding.

It is well known that the cold resistance of steel is determined by the state of the grain boundaries, the presence of interstitial impurities and their interaction with dislocations, the precipitation of excess phases, and the grain size. From this point of view, all technological conversions, i.e., smelting, hot-plastic treatment, and heat processing, are important for steels of all strength categories.

Sulfur, phosphorus, and interstitial impurities, such as, hydrogen, oxygen, and nitrogen, have a harshly negative effect on the cold resistance of steel. With an increase in the sulfur content, the number of sulfide inclusions, which play the role of stress concentrators, the anisotropy of impact toughness increase [28]. The embrittling effect of phosphorus has implications for the enrichment of grain boundaries within it due to strong segregation, as well as the formation of stress concentrators, i.e., the phosphide eutectic composition. Nitrogen also embrittles steel via blocking dislocations. The most dangerous consequences of the presence of nitrogen in low-alloy steel are a decrease in impact toughness, an increase in the cold-brittleness threshold, and the manifestation of a tendency of strain aging; therefore, its content is limited to 0.008 wt %. The steel quality with respect to the content of harmful impurities (sulfur and phosphorus), nonmetallic inclusions, and gases is also important to ensure good weldability, including weldability at low temperatures (down to –20°C) of the surrounding air during the welding of hull structures in open docks.

The following conditions were implemented in metallurgical technologies during the creation of the latest generation of cold-resistant steels: a decrease in the carbon content; a decrease in the degree of segregation in a continuously cast slab; an increase in steel purity from harmful impurities (sulfur and phosphorus) and impurities of nonferrous metals, gases, and nonmetallic inclusions; and the granulation of sulfide inclusions.

Since a simultaneous increase in the strength and resistance to brittle fractures can only be achieved with strengthening via the refinement of structural elements, it is necessary to use technological processes that ensure the formation of highly dispersed structures to create cold-resistant steels. The most significant increase in the dispersion of the structure at the microscopic and mesoscopic levels can be achieved during plastic deformation via an increase in the dislocation density, self-organization of dislocation structures, and the formation of disoriented microscopic regions [29–32].

In accordance with the laws established by Hall and Petch and developed by Rybin, Trefilov, Mil'man, Firstov, and other scientists, the increase in the yield point $\sigma_{0,2}$ is determined by the contribution of solid-solution hardening by carbon atoms (σ_C) and substitutional alloying elements (σ_s) and the contributions from various fine-structure components, such as, particles (σ_p), dislocations (σ_d), and large-angle (σ_{LAB}) and small-angle (σ_{SAB}) boundaries; the decrease in the brittle-to-ductile transition temperature T_{bd} of low-carbon steel depends on the size of the structure component at the microscopic and submicroscopic levels:

$$\sigma_{0.2} = \sigma_0 + \sigma_C + \sigma_s + \sigma_{SAB} + \sigma_{LAB} + \sigma_p + \sigma_d; \quad (5)$$

$$T_{bd} = A - B \ln d^{-1/2}, \quad (6)$$

where σ_0 is the friction stress of the lattice, d is the size of the effective grain or subgrain, and A and B are the metal constants.

Interfaces and dislocations make the main contribution to the strengthening of low-carbon steels with cost-effective alloying; their density can be adjusted with controlled plastic-deformation methods. Grain refinement via microalloying with vanadium, niobium, and/or titanium and the addition of aluminum is a prerequisite.

The ductile-to-brittle transition is associated with a decrease in the energy intensity of the fracture process and is caused by an increase in the yield point at low temperatures, as well as by the transformation of the fracture micromechanism. Since the plasticity margin in the metal becomes insufficient due to the presence of various defects, the conditions for local microcleavage are fulfilled.

During brittle fracture, a brittle fracture spot appears first. It is a microcrack that gradually develops to a critical size. An avalanche-like propagation of cracks in the structure is observed in the second stage, which is the most dangerous one. The capacity for plastic deformation at the minimal operating temperatures and the inhibition of microcracks, which are always present in the structure, are the most important steel properties with respect to its performance at low temperatures.

Zav'yalov, Kroshkin, Sakhin, Gorynin, and other scientists have demonstrated in studies that steel with a fibrous structure of the fracture after the destruction of large samples is much less prone than steel with a crystalline fracture to brittle failure and the development of cracks. It was found [33] that the number of deformed grain layers is a few dozen when the fracture is completely fibrous at temperatures above the upper brittleness threshold; the number of deformed grain layers decreases sharply at the lower brittleness-threshold temperature when the fracture becomes completely crystalline. This causes a considerable difference in the mechanical energy spent on grain deformation and, consequently, a sharp difference in the resistance of steel to the initiation and development of cracks. In this regard, the fibrous structure of the fracture at the lowest possible operating temperatures is a reliable criterion for the performance of structural steel. Therefore, low-temperature static bending tests on large samples to determine the ductile-to-brittle transition is a prerequisite for the assessment of the quality of shipbuilding steel in Russia, and room-temperature tests with estimation of the fraction of the fibrous component in the fracture is a prerequisite for acceptance testing (standardized in the range of 65–90% for shipbuilding steels, depending on the strength category).

It should be noted that the effectiveness of methods to increase the cold resistance of a material, such as structural refinement, alloying with nickel or other substitutional atoms, and a decrease in the content of S, P, and interstitial impurities (N and O), in the overwhelming majority of studies is controlled solely by changes in the temperature dependences of the impact toughness. In acceptance tests, the uniformity of the metal quality and the reproducibility of the technology are essentially assessed based on the impact work [34].

There are very few studies in which the influence of various factors on the crack-resistance parameters or the determination of the critical temperatures of the ductile-to-brittle transition are systematically analyzed with other techniques that involve full-thickness test samples. The physical concepts of the mechanisms to change the viscosity and cold resistance under the influence of various factors are also insufficiently developed.

The best-known formulation of the local criterion for brittle failure is expressed as follows (schemes of Davidenkov, Kopel'man, and Meshkov [2, 6]):

$$\sigma_{\max} \geq S_c, \tag{7}$$

where $\sigma_{\max}$ is the maximum principal stress and S_c is the true pull-off stress, which is a temperature-independent characteristic of the material.

In the substantiation of this criterion, the S_c value is assumed to correspond to the stress of the propagation of the so-called "Griffith" microcrack, which forms in the bulk of the grain through the structural barrier into the neighboring grain. In this scheme, the brittle-failure condition is determined by the grain size (or the size of a substructure component), which is the only structure-sensitive factor determining the effective length of the dislocation accumulation in front of the structural barrier.

A more complicated fracture criterion that takes into account the critical conditions for the nucleation, origination, and propagation of a microcrack was proposed and experimentally confirmed in [35, 36]. The failure conditions formulated here contain a number of structure-sensitive parameters.

Metallurgical approaches to the creation of cold-resistant structural steels are based on the relationships—which have been studied in sufficient detail over the past 10–15 years—between alloying, phase-transition kinetics, heat-strain regimes of hot-plastic deformation, and the parameters of accelerated cooling, quenching, and high-temperature tempering, on the one hand, and the formed structure, mechanical properties, and performance characteristics, on the other. With respect to structural steels, these issues were studied at the Prometei Central Research Institute of Structural Materials (a sectoral institute of shipbuilding since the 1970s), Bardin Central Research Institute of Ferrous Metallurgy, Institute of Metal Physics of the Ural Branch of the Russian Academy of Sciences, Mel'nikov Central Research and Design Institute of Building Constructions, Baikov Institute of Metallurgy of the Russian Academy of Sciences, and some other teams [37–43].

As a result, alloying principles were proposed and technologies were developed for the production of rolled sheets corresponding to shipbuilding strength categories of 235–690 MPa with a thickness of up to 70 mm [43, 44] and cold-resistant steels from pipe categories with a yield point of at least 500–690 MPa [45] and a thickness of up to 35 mm thick; they are described below.

The presence of carbon, a strengthening element that forms an interstitial solid solution with ferrite, in steel leads to a decrease in dislocation mobility and, as a result, to increases in the strength of the steel and in its tendency of brittle fracture. Carbon lowers the impact strength and raises the cold-brittleness threshold [46]. The quenching ability of steel in the HAZ increases upon an increase in the carbon content, which manifests itself in an increase in the hardness. This is associated with the danger of the appearance of cold cracks in welded joints.

Due to the need to ensure high weldability, the maximum carbon content in cold-resistant steels for structures operating in the Arctic region should not exceed 0.10–0.12 wt % and the content of alloying elements should be as low as possible. For rolled-metal sheets produced via thermomechanical processing, it is advisable to limit the carbon content to 0.05–0.07 wt % for manganese and manganese–nickel steels and to 0.07–0.10% for chromium–nickel–molybdenum steels; at the same time, microalloying with vanadium, niobium, and titanium—individually or in combination—to a total content that should not exceed 0.12 wt % and deoxidation with aluminum for grain refinement are prerequisite conditions.

The limitation on the carbon content creates certain difficulties in the preparation of heavy-duty structural steel and requires compensation for the decrease in the contribution of carbon to hardening via appropriate alloying, the use of controlled rolling, thermomechanical processing, etc.

Apparently, the role of the ferrite component in providing strength and other properties to steel increases with a decrease in the carbon content, since the content of carbide phase decreases and the amount of carbide-forming elements dissolved in ferrite increases.

In this regard, the alloying of steel is of particular importance. It should ensure the obtainment of the necessary properties of ferrite, such as a high strength, ductility, impact toughness, resistance to brittle fracture, etc. [47].

The alloying of steel with manganese, copper, and silicon slightly increases the hardenability, as well as the ferrite strength. However, the effectiveness of the use of manganese as an alloying element for high-strength, low-carbon steel is not promising for the following reasons [33]: a sharp increase in sensitivity to overheating, even at a manganese content of about 1 wt %; significant development of the tendency of temper brittleness at a manganese concentration of more than 1.5 wt % in combination with the inevitable presence of phosphorus in steel; and a high interaction of manganese with the furnace lining and formation—in contrast to the well-known desulfurizing effect of small amounts of manganese additives—of a large number of nonmetallic inclusions that contribute to the development of anisotropy of mechanical properties. Silicon causes significant distortion of the α-iron crystal lattice, increases the resistance to dislocation movement and the hardness of ferrite, and enhances the tendency of brittle fracture.

All of the main alloying substituting elements for shipbuilding steels, i.e., Cr, Ni, Mo, V, Mn, and Ti, belong to the α-transition metals of periods 4 and 5 of the Mendeleev periodic system of elements. Grigorovich considered the possibility of constructing a theory that would explain the metal structure based on the model of itinerant valence electrons that interact with the crystal lattice and the formation of directed bonds by subvalence electrons of the outer ion shells [48]. Shared electrons carry out the metallic interaction of atoms in the first coordination sphere in the $\langle 100 \rangle$ directions. The metallic interaction of ions with shared electrons, which determines the plasticity and toughness of steel, is superimposed on more rigid covalent bonds directed along the edges of the cube and can lead to embrittlement of the metal in the case of a significant overlap of the orbitals. Multivalent elements with orthogonal configurations of *d* or *p* electrons in solid solutions (molybdenum, vanadium, chromium, silicon, etc.) reduce the resistance to brittle fracture due to an increase in the covalent component of the interatomic bond [48, 49].

The authors of [50] noted that high plasticity at low deformation temperatures can be obtained when covalent bonds in the bcc lattice are suppressed via doping with a certain element and the formation of a substructure in the crystal subsystem. The covalent component can be substantially weakened by the introduction of elements with a spherical shell configuration of valence electrons.

Preference should primarily be given to elements with a spherical configuration of valence electrons during the alloying of cold-resistant steels. Nickel is among the most promising elements discussed above. Without decreasing the electron concentration during the transition to a solid solution with iron, nickel replaces rigid covalent exchange bonds with metallic ones. This apparently determines its well-known effect as an element that increases resistance to brittle fractures. Moreover, the effect of nickel is also explained by the fact that it weakens the interaction of dislocations with interstitial atoms and the resistance of the crystal lattice to the movement of free dislocations, which leads to increases in the plasticity and crack resistance of steel [51]. The nickel content does not exceed 0.8 wt % in steels with an enhanced strength and can reach 2–3 wt % in high-strength, cold-resistant shipbuilding steels.

In the case of the dissolution of several elements in iron, their action does not add up, as a rule; the action of the elements is even more complicated when the transformations in a solid solution are also determined by the possibility of the formation of a carbide phase and the nature of the phase.

It can be assumed that silicon, manganese, and nickel are included in the composition of ferrite in low-carbon, low-alloy steel under improvement. Copper is slightly soluble in the ferrite, but its solubility increases in the presence of nickel [52]. As a rule, molybdenum and chromium are partly included in the carbide phase and alloy ferrite, whereas niobium, vanadium, and titanium are almost entirely bound in special carbides. Molybdenum effectively increases the stability of the carbide phase and, consequently, the tempering resistance of steel. The most beneficial effect of molybdenum is a weakening of the temper brittleness in the improvement of structural steels. To ensure good weldability, the content of alloying elements (primarily, carbon and chromium) in steel must be limited.

The influence of microalloying additives is mainly manifested during the formation of interstitial or substitutional solutions in the solid state, in their effect on the degree of dispersion of grains and nonmetallic inclusions, the structure of grain boundaries, and the fine structure, and in the neutralization of the influence of harmful impurities.

As microalloying elements, vanadium and niobium are quite similar in their influence and form the same type of carbides and nitrides. At the same time, the niobium ability to form carbide is higher than that of vanadium.

Niobium carbonitrides dissolve in austenite in much larger quantities and at lower temperatures than vanadium carbides. Niobium has a favorable effect on the steel properties, since it promotes the formation of fine grains in rolled products via the suppression of austenite recrystallization before the $\gamma \rightarrow \alpha$ transformation [53–57].

Vanadium carbonitride particles are more dispersed, since they precipitate at a relatively low temperature and cause the hardening of microalloyed steel due to precipitation hardening. The addition of vanadium in the amount of 0.05–0.20 wt % increases the steel strength properties by 10–13%, both in the hot rolled state and after quenching and tempering [58–61]. Vanadium is a very effective element for the inhibition of static recrystallization.

Significant strengthening can be achieved with the addition of relatively small amounts of microalloying elements when the following two factors act simultaneously: grain refinement and precipitation hardening due to extremely small and evenly distributed particles of the secondary phase [58, 61].

To obtain the required properties of cold-resistant steels, complex alloying within the limits allowed by the RMRS requirements is usually applied with allowance for the selected manufacturing technology.

The principles of the alloying of cold-resistant, shipbuilding, low-alloy steels with a yield point of more than 235–460 MPa are as follows:

— a decrease in the carbon content (to 0.05–0.07 wt %), especially with the use of thermomechanical processing (TMP), to ensure high plasticity, viscosity, and cold resistance, and a decrease in the volume fraction of the carbide phase;

— the use of a manganese alloying composition to create steels with a guaranteed yield point of 235–315 MPa and a manganese–nickel alloying composition to create steels with a guaranteed yield point of 355–460 MPa;

— alloying with manganese added in amounts of 0.6–1.6 wt % to manganese steels and 0.9–1.6 wt % to manganese–nickel steels;

— the addition of small amounts of nickel (0.10–0.40 wt %) to steels with a ferrite–pearlite structure to increase the toughness and cold resistance;

— the addition of small amounts of nickel and copper (up to 0.6–1.0 wt % in total) and molybdenum (up to 0.10 wt %) to steel with a ferrite–bainite structure to ensure a bainite fraction of at least 30–50 wt %, to increase the toughness and cold resistance to prevent grain growth and the formation of ferrite along grain boundaries in the HAZ of the welded joint;

— microalloying with vanadium and niobium in amounts of less than 0.03–0.06 wt % of each element and, less often, with titanium in an amount of less than 0.012–0.015 wt %, which makes it possible to control the grain size during heating and hot-plastic deformation, during self-tempering after rapid cooling in the process of TMP, and during welding;

— the obtainment of steel purity with respect to sulfur (no more than 0.005 wt %), phosphorus (no more than 0.010 wt %), nitrogen (no more than 0.008 wt %), and nonmetallic inclusions (no more than entities in accordance with GOST 1778 (State Standard of the former USSR)).

In high-strength, economically alloyed, cold-resistant steels (with a yield point of 500–960 MPa) intended for operation in the Arctic region, the carbon content should be reduced to $C \leq 0.12$ wt % and below 0.10 wt % in some cases to exclude brittle failures and to increase weldability during construction and installation. Complex alloying with nickel, chromium, copper, and molybdenum (with manganese additions) and microalloying should be used to compensate for the hardening effect of carbon. In this case, the chemical composition of the steel should guarantee the formation of a bainitic or bainitic–martensitic structure of high-strength steel in a fairly wide range of cooling rates, which ensure the absence of diffusion ferrite–pearlite transformation products throughout the thickness of the rolled sheet (which is important for thick rolled-sheet products).

From this point of view, it is necessary to take into account the effect of hot-plastic deformation in the austenitic region on phase transformations, since modern, high-strength, cold-resistant, economically alloyed steels are produced mainly via quenching from hot rolling in combination with high-temperature tempering (QHR + T). For the most critical structures, it is preferable to use hardening via furnace heating with tempering.

To date, Russia has developed a series of economically alloyed, cold-resistant steels with the Arc index with guaranteed yield points of 355, 390, 420, 460, 500, 620, 690, and 750 MPa. Steels with guaranteed yield points of 890 and 960 MPa are being developed in accordance with the requirements of Rules of the RMRS for the design and construction of ice-navigation vessels, research vessels, and icebreakers, including those with enhanced capacity, as well as offshore drilling platforms of various types.

In many respects, this was facilitated by a number of significant technical innovations that have been implemented in the domestic metallurgical industry over the past 10–20 years and have made it possible to improve the technology for the production of cold-resistant steels of the shipbuilding, pipe, and construction grades. This primarily concerns the smelting and rolling technologies.

PRODUCTION TECHNOLOGY FOR ROLLED SHEETS AND THEIR STRUCTURE AND PROPERTIES

Cold-resistant steels are smelted in converter or electric-arc furnaces via deoxidation, vacuum degassing, and modification with calcium and other elements.

For a sharp reduction in harmful impurities and gases in the metal, the modification of nonmetallic inclusions, and the protection of liquid metal from secondary oxidation during the casting process, the off-furnace treatment makes it possible to obtain sheet steels with a high level of ductility in the direction of the sheet thickness and ensures the absence of layered damage in the welding of T-joints.

As established in [62], the impact work, strength characteristics, and elongation under uniaxial tension of samples of a shipbuilding steel sheet in the direction of thickness are influenced by the morphology and distribution of nonmetallic inclusions of a metallurgical origin. Large inclusions and their agglomerates with a length of more than 150 μm facilitate the fracture process and lead to the appearance of splits and a decrease in the level of impact work and plasticity during tensile tests, including those performed in the Z direction.

The presence of chemical heterogeneity in the sheet structure due to insufficient elaboration of the structure in the process of hot-plastic deformation during TMP and the inheritance of axial looseness and chemical heterogeneity from the initial slab are also the reasons for a decrease in the impact work and relative contraction of low-carbon, low-alloy steel in the direction of the sheet thickness.

Undesirable splits in fractures of full-thickness samples during static and dynamic tests appear due to the presence of a layered, inhomogeneous structure and extended interfaces in rolled-metal sheets [63, 64]. Such boundaries in ferritic–pearlitic steels are represented by the boundaries between ferrite and pearlite colonies elongated in the direction of rolling and are represented in ferritic–bainitic steels by the boundaries between ferrite strips and elongated regions of lath bainite of the Widmanstätten structure type. Cracks originating from the interfaces can propagate both along the boundaries and along the body of ferrite grains adjacent to such boundaries.

Changes in the regime of thermomechanical processing of steel with ferrite–pearlite and ferrite–bainite structures to refine grain during dynamic and primary static recrystallization in the γ region makes it possible to obtain a dense, viscous fracture of samples without significant splits during static and dynamic tests.

Thus, it is important to take into account not only the ratio of structural components but also their morphological similarity, which contributes to the formation of a mixture of structural components with approximately the same dimensional and morphological characteristics in the creation of cold-resistant, shipbuilding steels. The following general principles were formulated for the selection of the optimal structure of low-carbon, cold-resistant weldable steel with high resistances to static, dynamic, and cyclic loads and corrosive environment:

— the obtainment of morphologically similar structural components (grain and subgrain sizes, density of dislocations, and dispersity of the carbide phase);

— an increase in the dispersity of the austenite structure and the transformed structure;

— exclusion of the formation of extended interfaces in steel with a mixed structure and extended lamellar carbides on them.

This problem is solved both via the selection of a chemical composition that makes it possible to vary the cooling rate in a fairly wide range without changes in the nature of the phase transitions and the successive refinement of the structure at each stage of the technological chain.

Regarding the nature of the structural response to the thermomechanical influence, the following four effects can be distinguished:

(1) the refinement of austenite grains due to dynamic recrystallization during hot-plastic deformation in the high-temperature region (above the temperature of recrystallization);

(2) the creation of a developed substructure in the austenite due to fragmentation and polygonization during plastic deformation at temperatures 100–250°C below the recrystallization-threshold temperature;

(3) the formation of dispersed finite structures due to phase transformation in a given region in the cooling process;

(4) the formation of a developed subgrain structure in the final (transformed) phase during plastic deformation below the temperature Ar_3 (for steels with a ferrite–pearlite structure).

The structural state of deformed austenite largely determines the processes of structural formation during the subsequent phase transition. It was established that the fraction of small-angle misorientations in the crystallographic spectrum increases in the case of deformation of the initial austenite, which can be explained by the fragmentation of austenite and the inheritance of small-angle misorientations of strain origin [65, 66]. Hot-plastic deformation in the fragmentation region of austenite leads to an increase in the number of dislocation boundaries in the internal volume of bainitic ferrite regions. The average sizes of structural components formed by such boundaries decrease from 2 to 0.5–0.6 μm after preliminary deformation with a strain degree of $\varepsilon = 30\%$ at a temperature of 900°C; the boundaries of bainite and martensite laths are curved, which prevents the formation of extended lamellar carbides.

To develop thermomechanical processing regimes that provide the maximum possible refinement of the grain and subgrain structures of low-carbon steels, detailed information on the effect of the size of the initial austenite grain and the degree of its fragmentation on the phase transition kinetics, the type and quantitative characteristics of the forming ferrite, ferrite–bainite, and bainite–martensite structures was obtained in previous studies [67, 68].

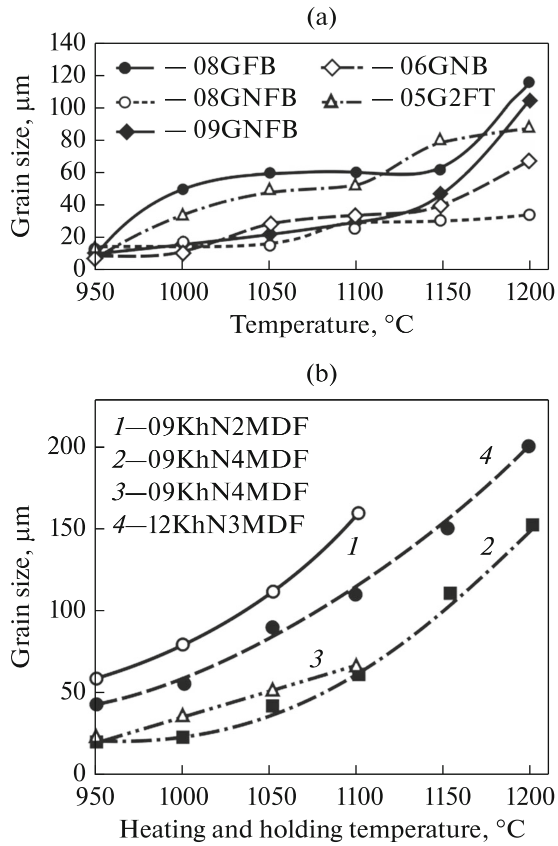

Fig. 1. Change in the size of the austenitic grain upon the heating of (a) low-alloy and (b) medium-alloy high-strength steels [69].

CHARACTERISTICS OF GRAIN GROWTH DURING HEATING AND DYNAMIC AND STATIC RECRYSTALLIZATION PROCESSES IN AUSTENITE OF COLD-RESISTANT STEELS ALLOYED WITH VARIOUS ELEMENTS

Microalloying elements have the main effect on the processes of austenite-grain growth during heating and the recrystallization processes in low-alloy and economically alloyed shipbuilding steels. Niobium added in an amount of 0.02 wt % effectively slows grain growth and ensures the production of fine-grained steel (nos. 9–11 according to GOST 5639, Fig. 1).

Large amounts of niobium are not as effective, but they increase the precipitation hardening. To obtain the same strengthening effect, the vanadium content should be two to three times higher than the niobium content.

Under typical rolling conditions, the addition of 0.03 wt % niobium to low-carbon steel completely suppresses austenite recrystallization at temperatures of about 950°C. Compared to vanadium microadditives, a smaller amount of niobium slows austenite recrystallization at higher rolling temperatures.

Microalloying with elements capable of forming carbonitrides, especially niobium and titanium, prevents dynamic and metadynamic recrystallization processes and helps to reduce their intensity [53]. This effect is already noted at temperatures of 1000–1100°C [70], but it is most pronounced at temperatures below 950°C. The processes of the precipitation of niobium and titanium carbonitrides are substantially accelerated under the influence of austenite deformation. The deformation-stimulated precipitation of carbonitride phases proceeds most strongly at 925–875°C. It is this range in which the recrystallization of steels microalloyed with niobium or titanium sharply slows down relative to the recrystallization rate in unalloyed steel.

In low-alloy manganese steel containing 0.03 wt % niobium, the formation of dynamically recrystallized austenite grains with an average diameter of 14–15 μm is observed at a deformation temperature of 1000°C for various strain degrees in the range of 20–80% (Fig. 2a).

After deformation with a strain degree of 20%, about 25% of unrecrystallized austenite grains are formed at a temperature of 950°C (Fig. 2b), and traces of dynamic recrystallization are completely absent at a temperature of 900°C (Fig. 2c). The completion of dynamic recrystallization at temperatures of 900–950°C is observed with a strain degree of no less than 80% (Fig. 2d), which is unattainable in the rolling process.

Table 5 shows the temperature ranges of recrystallization in low-alloy manganese steel microalloyed with niobium at deformation rates of 15–20% per pass that were achieved in the industrial environment.

In chromium–nickel–copper–molybdenum steel, dynamic and, consequently, metadynamic recrystallization processes are unlikely under conditions of fractional hot rolling, and the main mechanism for the formation of a fine-grained austenite structure is static recrystallization in the pauses between successive deformations [55, 72].

The replacement of vanadium with niobium in high-strength chromium–nickel–copper–molybde-

Table 5. Temperature ranges of recrystallization processes of low-alloy manganese steel microalloyed with niobium [71]

Recrystallization type		
Dynamic	Metadynamic	Static
Not less than 1000°C	1000–950°C	900°C, after holding for at least 5 min

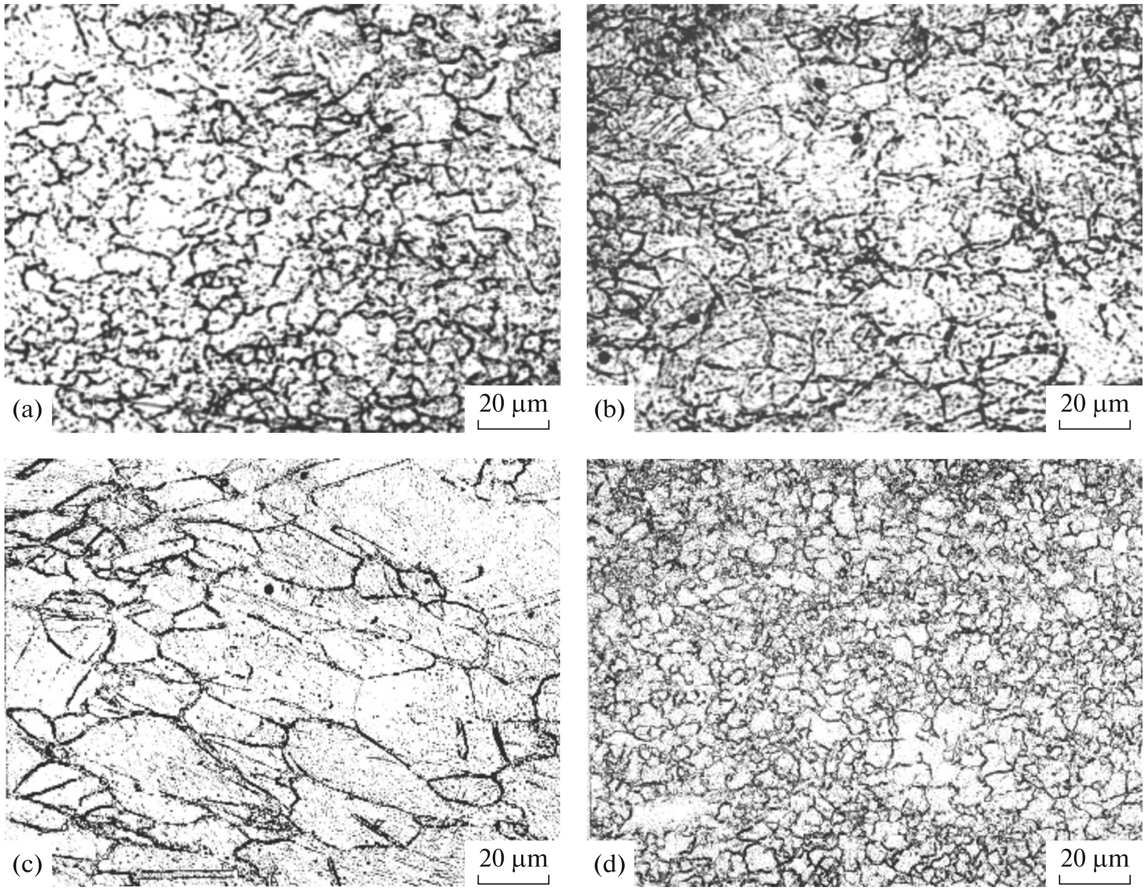

Fig. 2. Size of the austenitic grain in low-alloy steel after deformation to a degree of 20% at temperatures of (a) 1000°C, (b) 950°C, and (c) 900°C and (g) to a degree of 80% at a temperature of 900°C without holding [71].

num steel gives rise to the production of a more uniform and finely dispersed austenite structure after hot rolling due to a number of positive effects: (a) a restriction of grain growth upon heating of the workpiece for rolling; (b) the suppression of dynamic recrystallization, which can be only partial in view of technological limitations and thereby leads to structural heterogeneity; (c) prevention of the growth of new grains after the completion of the primary static recrystallization in the pauses between rolling actions in the high-temperature (rough) rolling stage; (d) expansion of the temperature range of fragmentation, which leads to the formation of new boundaries in austenite grains after the completion of the recrystallization stage.

At the same time, the regimes of fractional rolling of economically alloyed steel should be corrected with allowance for the retardation of recrystallization during microalloying with niobium in order to maintain the efficiency of multiple static recrystallization in the refinement of austenite grains. This requires a corresponding increase in the fraction of rolling action on the workpiece in the upper part of the rolling temperature range ($T > 1050°C$) [55].

The process of fragmentation at temperatures of $T < T_{sr} \approx 850°C$ is also capable of the formation of a finely dispersed austenite structure in chromium–nickel–copper–molybdenum steel with large-angle misorientations between internal grain microregions, but this requires an accumulation of true deformation of about unity (63% of sheet thinning), which is associated with serious technological limitations at the indicated temperatures. To overcome such limitations, it is necessary to increase the temperature T_{sr} of the onset of static recrystallization via the introduction of microalloying elements.

Thus, the microalloying elements in shipbuilding steel largely determine the processes occurring during hot-plastic deformation prior to rapid cooling.

Modern rolling equipment allows the use of automated hot-plastic deformation technologies due to the possibility of high-speed rolling and high-precision control of the deformation temperature scheme [73, 74], the degree and fractionality of rolling actions [75], and the duration of pauses between deformations [72].

Thermal-deformation treatment with controlled cooling to a given temperature (RC) or to room temperature (QHR) is the most effective method to simultaneous increase all of the performance characteristics of structural steel. It is carried out in powerful rolling mills equipped with units for rapid cooling. The treatment should take into account the described physical processes in the high-temperature and low-temperature regions during hot deformation [29, 54, 72, 76–90], as well as the phase transformations during cooling, which ensure the formation of the final transformed structure with a given degree of dispersity and morphology.

CHARACTERISTICS OF PHASE TRANSFORMATION IN LOW-CARBON, LOW-ALLOY, AND ECONOMICALLY ALLOYED STEELS UPON CONTINUOUS COOLING

By virtue of their alloying, shipbuilding steels in the state of delivery (thermomechanical processing and quenching from hot rolling with tempering) are characterized by a mixed ferrite–pearlite, ferrite–bainite, or bainite–martensite structures. The hardenability of steel is also determined by the sheet thickness and cooling rate.

Alloying elements (Mn, Cr, Cu, Ni, and Mo) increase the stability of supercooled austenite and lead to the formation of low-temperature transformation products of austenite with a higher dispersity. The products of diffusion transformation reduce the strength characteristics and the resistance of high-strength steel to brittle fractures; therefore, the alloying of steel with a guaranteed yield point of 500 MPa or higher should ensure that the shear transformation takes place in the intermediate or martensitic regions of sheet products of the required thickness during cooling. Strong carbide-forming elements (vanadium, niobium, and molybdenum) increase the hardenability of the steel if they are dissolved in austenite and decrease it if they are bound into carbides or carbonitrides.

Niobium and vanadium have an effect on the kinetics of the $\gamma \rightarrow \alpha$ transformation. On the one hand, these elements inhibit the growth of austenite grains during heating, due to which they contribute to a decrease in the thermal stability in the temperature range of ferrite–pearlite transformation, shift the bainite region towards higher cooling rates on the thermokinetic diagrams of austenite decomposition during continuous cooling, and increase the critical points of the beginning and the end of austenite decomposition. On the other hand, they are present in the form of particles of an excess phase and can serve as additional sites for the nucleation of a new phase and contribute to a decrease in austenite stability [90].

The nucleation of ferrite grains at the initial austenite boundaries is substantially accelerated in deformed austenite [91]. The analysis carried out in [92] showed that this effect cannot be explained simply by an increase in the number of potential nucleation sites at the boundaries. The nature of transformation curves can be reproduced only with the assumption of a significant decrease in the energy barrier for the nucleation of a new phase. It turns out that an increase in the driving force of the transformation due to the stored energy of the deformed austenite substructure is too small to ensure the observed acceleration of the transformation [92]. Thus, it must be assumed that the decrease in the nucleation barrier is mainly associated with an increase in the energy of the boundaries themselves due to the increase in the degree of their defectiveness and with the formation of a near-boundary deformed substructure.

For steels with a higher content of alloying elements, a more substantial increase in the critical point of the beginning of transformation is caused by increases in the effective surface of the austenite interface and the number of ferrite nucleation centers. The resistance of austenite to pearlite transformation, which is observed at higher cooling rates, decreases in the deformed state as compared to the undeformed state.

Since the transformation of austenite in the second stage (bainitic transformation) has traits of both diffusion and diffusionless transformations, the deformation of austenite has a dual effect on it. On the one hand, the temperatures of the beginning and end of the bainitic transformation increase; on the other hand, the bainite region in the thermokinetic diagram expands to the right toward lower cooling rates.

Austenite deformation has an effect on the transformation kinetics, because the nucleation of a new phase at grain boundaries is accelerated and additional nucleation sites appear inside the grain [92, 93] (Fig. 3).

Nonetheless, it is necessary to take into account a certain inhibitory effect of deformation, since the boundaries of fragments and subgrains are the barriers to grain growth. In the fragmented austenite structure, regions of large-angle boundaries of strain origin appear at the sites of the most energetically favorable nucleation [29]. The rate of intragranular nucleation of primary bainite components is proportional to the specific surface area of such regions. In turn, this value is proportional to the specific surface area of the subboundaries and the fraction of boundaries of strain origin with misorientation angles of more than 8°.

The expansion of the range of bainitic transformation to lower cooling rates, which makes it possible to obtain a bainite structure that leads to an increase in the strength characteristics in a wider range of technological modes and, most importantly, sheet thicknesses, can be considered a positive effect of hot-plastic deformation in the austenitic state.

The low contents of carbon and microalloying elements that increase the stability of austenite lead to the formation of granular bainite in steels with a minimum content of bainite of lath morphology, which

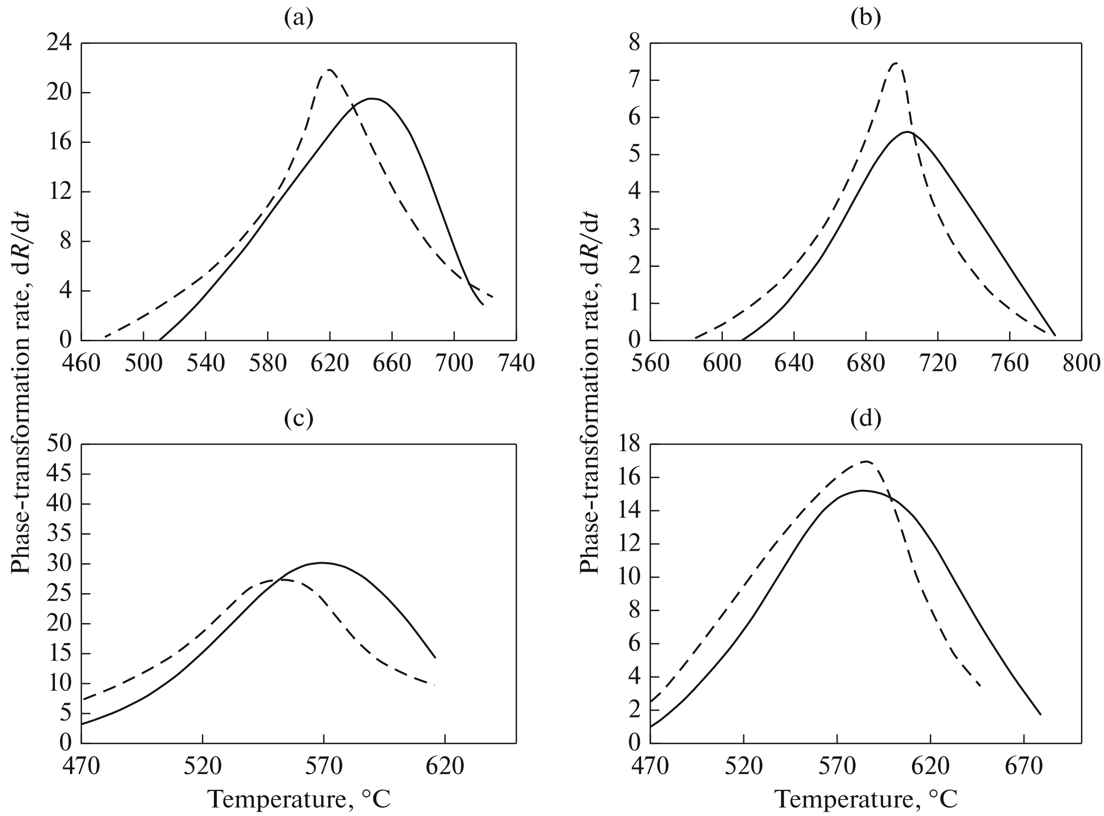

Fig. 3. Influence of preliminary hot deformation on the phase-transformation rate of microalloyed steels (a, b) 06GNFB and (c, d) 09KhN2MD with (a) $v_{cool} = 25°C/s$, (b) $v_{cool} = 5°C/s$, (c) $v_{cool} = 50°C/s$, and (d) $v_{cool} = 20°C/s$ [93] without deformation (solid line) and with deformation in the austenitic region (dashed line).

reduces the resistance to brittle fractures due to the formation of extended lamellar carbides along the lath boundaries.

The structure of deformed austenite is inherited during bainitic transformation, and misorientations at the boundaries of fragments in bainitic regions are not only inherited but also increase [93].

Low-alloy manganese and manganese–nickel steels with a guaranteed yield point of 315–460 MPa. A comprehensive study of the phase and structural transformations of low-alloy shipbuilding steels with different carbon equivalents in the range of $C_{eq} = 0.20–0.45\%$ (due to variation of the contents of carbon and the main alloying elements, i.e., manganese, nickel, copper and molybdenum) with cooling rates of 5 to 30°C/s, which are typical of the cooling of the surface and the middle part of rolled-metal sheets with a thickness of 100 mm, was carried out in an industrial environment. The formation of a ferrite–pearlite–bainite structure at $C_{eq} = 0.20–0.25\%$ (for steel of the strength category of 315 MPa, Fig. 4a) and a ferrite–bainite structure at $C_{eq} = 0.29–0.40\%$ (for steels of the strength categories of 355–460 MPa, Figs. 4b and 4c) can be ensured for this entire range.

With an increase in the C_{eq} values, the fraction of ferrite decreases and the bainite component increases simultaneously, which results in an increase in the strength characteristics. A further increase in the C_{eq} values to 0.43–0.45% in low-alloy steel can lead to the formation of a predominantly bainitic structure containing bainites of various morphology with cooling rates of at least 5°C/s (Fig. 4d).

Economically alloyed chromium–nickel–copper–molybdenum steels with a guaranteed yield point of 500–960 MPa. A bainite–martensite or martensite structure is formed in steels at $C_{eq} = 0.51–0.59\%$ with a total (Ni + Cu + Mo) content of around 2.5–3.0 wt % and a Cr content of 0.5–0.7% upon cooling at rates of 5°C/s and higher, which are typical of the quenching of sheets with a thickness of up to 50 mm (Figs. 5b and 5c). A decrease in the total content of elements that stabilize austenite to 2% makes it impossible to ensure the required hardenability of steel in sheets with larger thicknesses (Fig. 5d). An increase in the chromium content to 1 wt % or more with a total (Ni + Cu + Mo) content of around 2.0–2.5 wt % promotes ferrite transformation upon cooling at rates of 5–10°C/s (Fig. 5a), which are typical of the cooling of the middle part of sheets with a thickness of no less than 30 mm [94].

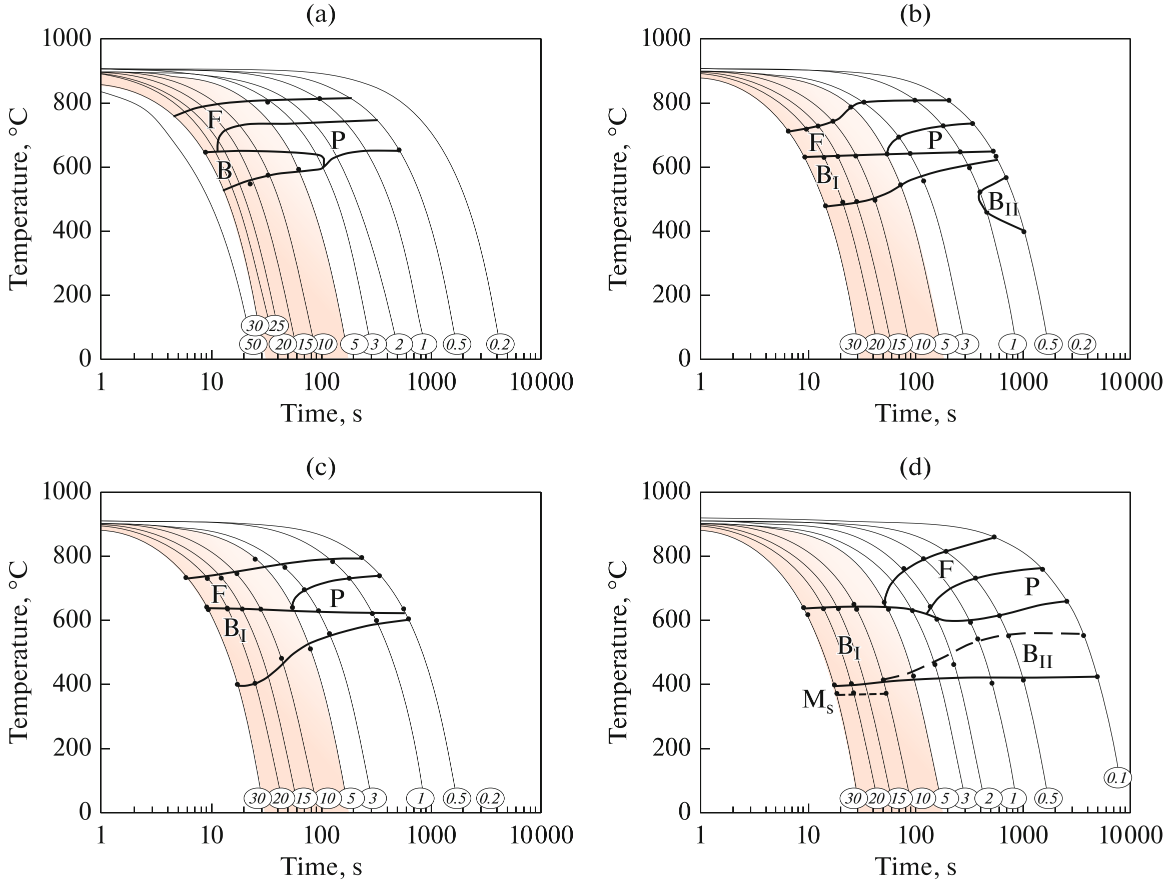

Fig. 4. Thermokinetic diagrams of the $\gamma \rightarrow \alpha$ transformation of fine-grained austenite of low-alloy steels with different carbon equivalents: (a) $C_{eq} = 0.21\%$; (b) $C_{eq} = 0.29\%$; (c) $C_{eq} = 0.38\%$; (d) $C_{eq} = 0.44\%$.

Controlling the structure-forming processes (grain growth, recrystallization, fragmentation, polygonization, and phase transformations) described above via variation of the temperature and deformation-rate conditions for the hot-rolling and rapid-cooling processes makes it possible not only to change the ratio and morphology of structural components but also to control the size of structural components.

IMPROVING THE HOT-PLASTIC DEFORMATION REGIMES

The obtainment of performance characteristics of cold-resistant, shipbuilding steels with various strengths with a reduced production cost is currently the most important issue in the development of technologies for the production of rolled-metal sheets made of cold-resistant steels.

In Russian metallurgical plants, two-stage thermomechanical processing with rapid cooling (TMP + RC) was chosen as a technological scheme for the production of low-alloy steels, and quenching from hot rolling with controlled hot-plastic deformation regimes and subsequent high-temperature tempering (QHR + T) was chosen as a technological scheme for the production of economically alloyed, high-strength steels.

Low-alloy steels of strength categories of 235–315 MPa. It was established that the final (finishing) hot-plastic deformation of sheet metal for the mass production of low-carbon, shipbuilding steels with a guaranteed yield point of 235–315 MPa should begin near and end slightly below the critical point Ar_3 (by 20–40°C). This makes it possible to form the maximum possible number of ferrite nuclei at the boundaries of austenite grains and to ensure the formation of a developed substructure in ferrite due to its insignificant cold-hardening and the completion of the polygonization process [93].

The deformation of fine-grained austenite in the region of its fragmentation leads to refinement of the ferrite grain (Fig. 6), and a decrease in the deformation temperature below the Ar_3 point promotes additional hardening due to an increase in the dislocation density (Fig. 7a) and small-angle boundaries of strain origin (Fig. 7b) [93].

Due to the use of rational technological processing schemes, the hardening effect of the smallest nano-

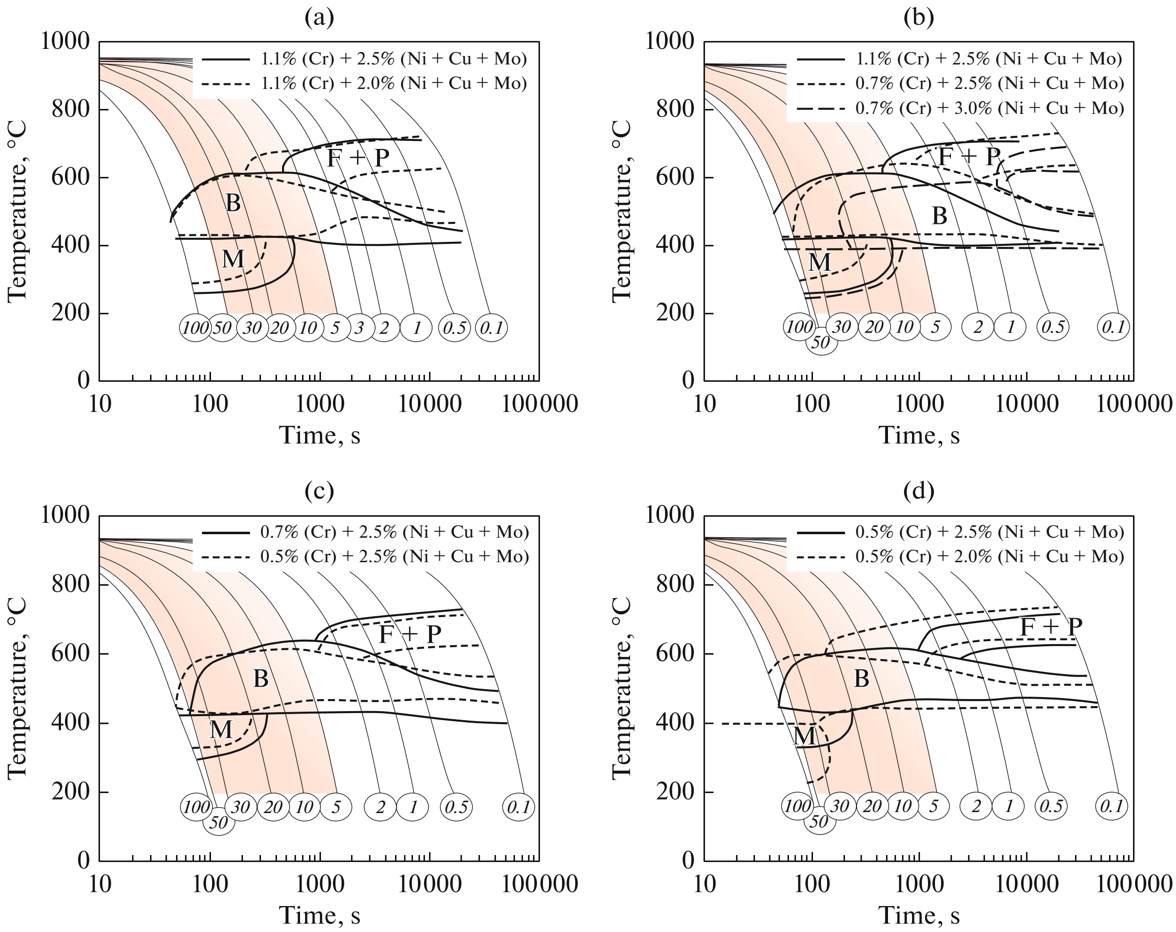

Fig. 5. Thermokinetic diagrams of the $\gamma \rightarrow \alpha$ transformation of fine-grained, unstrained austenite of economically alloyed, high-strength steels with different chromium contents and total contents of nickel, copper and molybdenum [94]: (a) 1.1% Cr with total (Ni + Cu + Mo) contents of 2.0 and 2.5%; (b) 0.7% Cr with total (Ni + Cu + Mo) contents of 2.5 and 3.5%; (c) 2.5 (Ni + Cu + Mo) with Cr contents of 0.5 and 0.7%; (d) 0.5% Cr with total (Ni + Cu + Mo) contents of 2.5 and 2.0%.

precipitates of complex carbonitrides of microalloying elements is used. Their size and bulk density can be controlled with adjustments of the temperature–strain regimes of rolling and the parameters of rapid cooling [93].

Due to the precipitation of nanodispersed particles (5–10 nm in size) of vanadium and niobium carbonitrides (Fig. 8), the yield point of the 05GFBT microalloyed manganese steel increases to 500–550 MPa, which is accompanied by simultaneous improvement of the impact toughness at test temperatures in the range from –60 to –80°C (Fig. 9). Vanadium was noted to have a positive effect on ferritic–pearlitic steels [93].

Thus, the combination of high strength and resistance to brittle fracture in rolled-metal sheets made of low-alloy, ferritic–pearlitic steels can be achieved via substantial refinement of ferrite grains as a result of the finishing stage of thermomechanical processing in a temperature range of Ar_3–(Ar_3 – (20–40°C)) and the application of complex microalloying with elements,

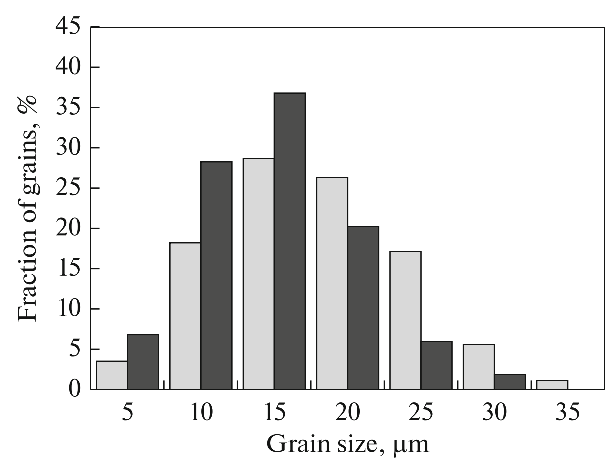

Fig. 6. Size distribution of ferrite grains (□) after the dynamic recrystallization of austenite and (■) after deformation of fine-grained austenite in the region of its fragmentation [93].

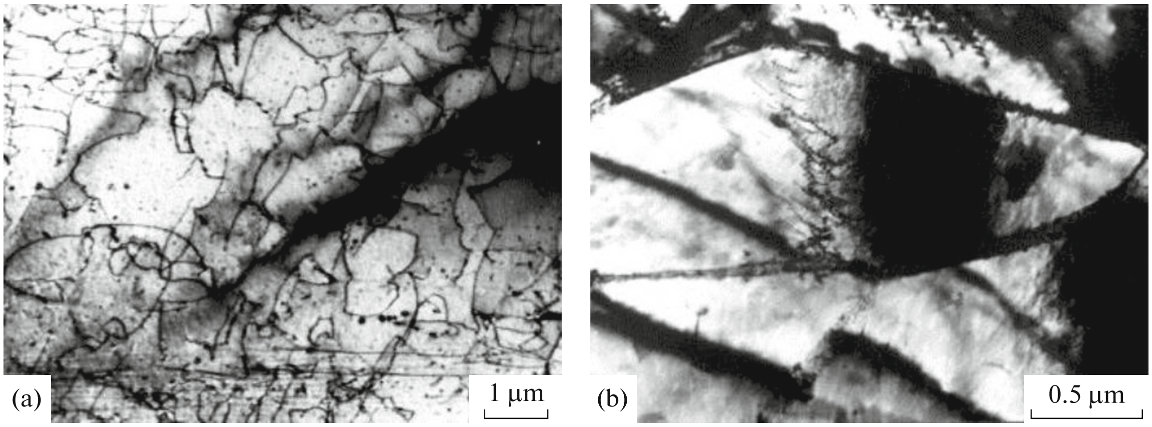

Fig. 7. Fine structure of the low-alloy ferrite steel 06GFB after thermomechanical processing with rapid cooling [93]: (a) dislocations; (b) SAB.

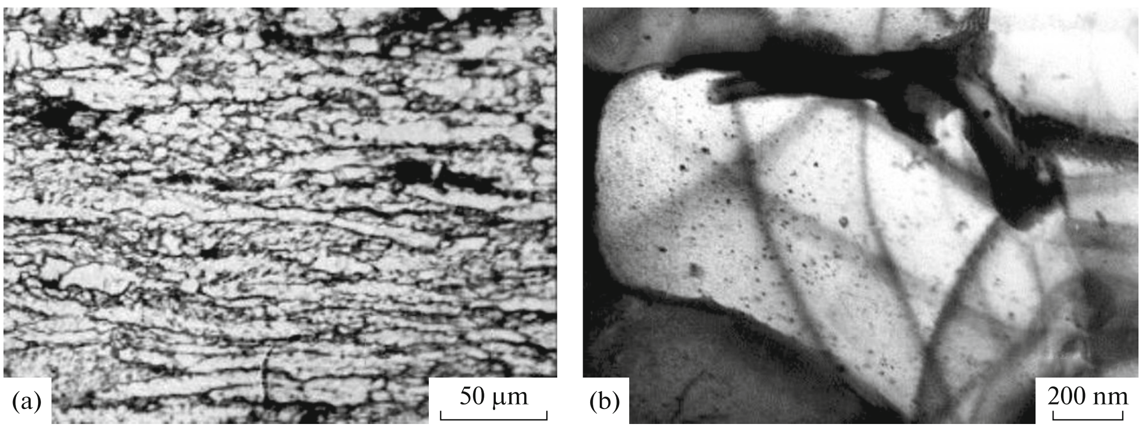

Fig. 8. Formation of (a) ultra-fine-grained structures and (b) carbide nanoprecipitates in low-alloy steel of the ferrite–pearlite class [93].

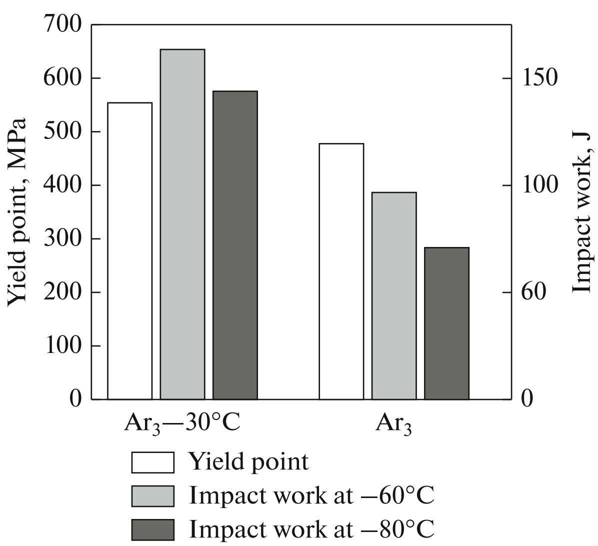

Fig. 9. Effect of the final rolling temperature on the yield point and impact-work values at low temperatures of the testing of low-alloy ferrite–pearlite steel [93].

such as V, Nb, Al, and Ti. At the same time, good steel weldability is ensured as a result of the low carbon content and limitation of the crack-resistance parameter P_{cm} to 0.16–0.18% during welding.

Low-alloy steels of strength categories of 355–390 MPa. Rolled sheets made of low-alloy cold-resistant steels with a guaranteed yield point of 355–390 MPa are also mainly produced with the technology of thermomechanical processing with rapid cooling (TMP + RC). The greatest problems with the obtainment of cold resistance arise in the manufacture of rolled-metal sheets with a thickness of more than 40 mm, for which the impact work at low test temperatures must be controlled on samples cut not only from the surface but also from the middle part across the sheet thickness.

Systemic studies for the creation of technologies to produce rolled-metal sheets from cold-resistant, shipbuilding steels for the Arctic region, including those with the Arc index, showed that traditional technological approaches do not allow the formation of a speci-

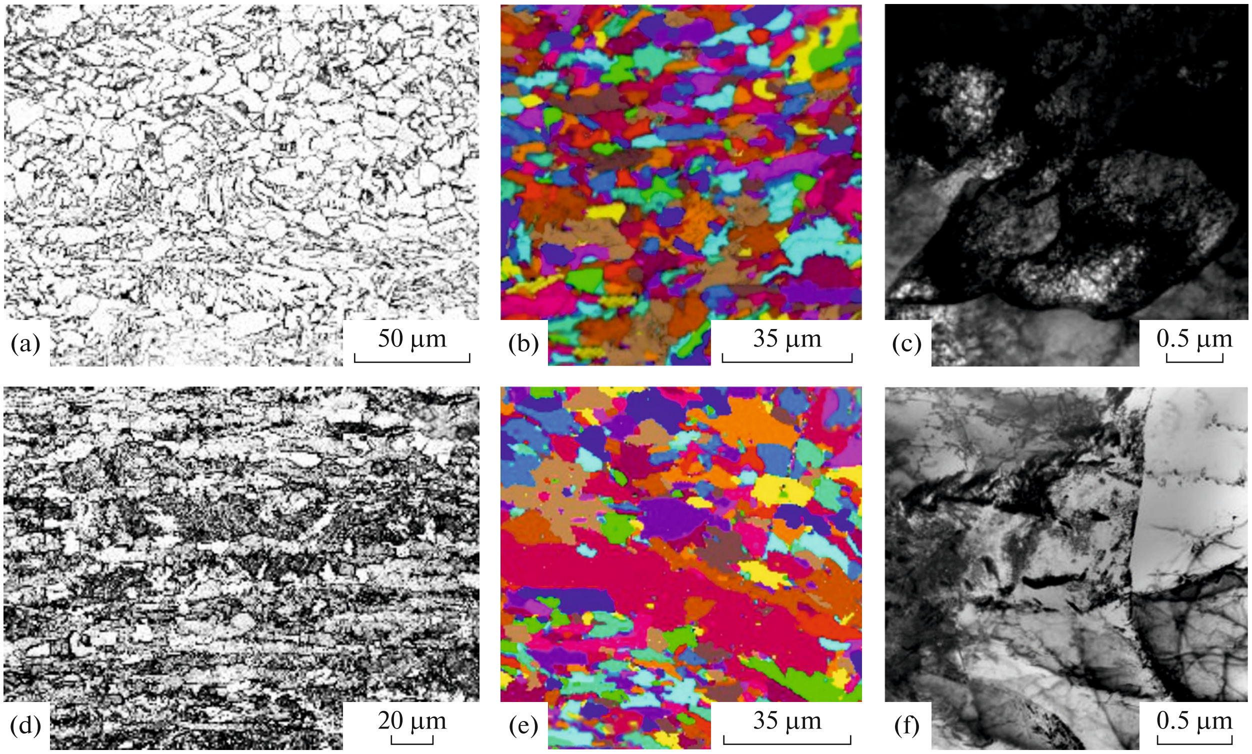

Fig. 10. Structure of a rolled sheet made of low-alloy steel belonging to a strength category of 355–390 MPa manufactured with the TMP + RC technology (a–c) in the optimal regimes and (d–f) with deviations from the optimal regimes.

fied, quasi-uniform, ferrite–bainite structure that would ensure guaranteed performance characteristics. The parameters of such a structure will be slightly different in the middle part of the sheet thickness.

It was established that it is advisable to conduct deformation at the rough (high-temperature) rolling stage in steps with maximum compression in an ascending scheme and with elevation of the temperature range with an increase in the carbon equivalent C_{eq} of low-alloy steel.

We proposed a method to control the morphology, ratio, and size of structural components in the final ferrite–bainite structure of low-alloy steel via the implementation of various temperature deformation schemes with controlled compressions at the finishing stage of rolling, which was carried out both in the isothermal and ascending regimes (with heating of the rolled workpiece due to high-speed plastic deformation) or a scheme of a descending temperature schedule in a narrow temperature range [73].

For low-alloy steels with a yield point of at least 355–390 MPa at the finishing (final) stage, the most effective use of an isothermal rolling scheme is achieved at a temperature 200–250°C below the threshold temperature of austenite recrystallization. This will give rise to the formation of the maximum possible number of nucleation sites for new grains to obtain a dispersed, ferrite–bainite structure.

Deformation at a constant temperature leads to a decrease in the proportion of lath bainite in the ferrite–bainite structure, an increase in the dispersity of structural components, and the formation of bainite with the granular morphology in combination with quasi-polygonal ferrite; this results in an increased stability of viscoplastic characteristics and resistance to brittle fracture.

Rolled-metal sheets made of low-alloy steels belonging to strength categories of 355–390 MPa are cooled to temperatures corresponding to the upper and middle parts of the bainitic transformation, at which austenite not decomposed by the diffusion mechanism at higher temperatures undergoes a transformation by the mixed mechanism.

The structure of 50-mm-thick rolled-metal sheets manufactured according to the described TMP + RC regimes is largely a mixture of structural components with similar morphological features (quasi-polygonal ferrite and bainite of granular morphology) (Figs. 10a and 10c). Quasi-polygonal ferrite is characterized by a developed subgrain structure and a high density of dislocations (1.1×10^{14} m^{-2}) that is close to the dislocation density in granular bainite ($2.0–2.2 \times 10^{14}$ m^{-2}) (Fig. 10c). The fraction of quasi-polygonal ferrite over the thickness of rolled sheets reaches 40–45%, and the fraction of bainite with the lath morphology, which is evenly distributed across the thin section, does not exceed 15%.

The results of electron backscattering diffraction (EBSD) analysis showed that the size of structural components is 15.2 μm on average at a given tolerance angle of 15° and 7.2 μm at 5° (Fig. 10b). This indicates that the regions of granular and lath bainite, as well as grains of quasi-polygonal ferrite, are substantially fragmented, which was confirmed by the study of the fine structure on a transmission electron microscope (Fig. 10c).

The formation of such a structure in low-alloy steels belonging to strength categories of 355–390 MPa guarantees the obtainment of high cold-resistance and crack-resistance characteristics at low test temperatures in 50-mm-thick steel sheets; the values of the impact work at test temperatures of –40 and –60°C are not lower than 150 J, and the temperatures of the ductile-to-brittle transition (T_{cb}) and zero plasticity (ZDT) are –15 and –65°C, respectively. The obtained values of the critical brittleness temperatures meet the requirements for new cold-resistant steels with the Arc40 index. At the same time, the mean displacement of the critical opening at the crack tip (CTOD) at a test temperature of –60°C is 1.26 mm and substantially exceeds the requirements even for steels with the Arc60 index (not less than 0.20 mm). The interfaces with misorientation angles of 5°–15°, along with misorientation boundaries with an angle of more than 15°, probably contribute to the resistance to crack initiation and propagation, which explains the high fracture toughness values of these kinds of steels.

The formation of a ferrite–bainite structure with pronounced structural anisotropy in the rolling direction and an anisomerous grain size (the size of ferrite grains vary in the range from 1 to 25 μm, Fig. 10e), as well as the presence of a significant fraction (up to 15%) of pearlite-like bainite (areas of polygonal ferrite with degenerate pearlite at boundaries, Figs. 10d and 10f) makes it possible to obtain the required strength characteristics but leads to significant decreases in the cold-resistance and crack-resistance values of the steel. The critical brittleness temperatures are T_{cb} = +5°C and ZDT = –40°C, and the mean displacement of the CTOD at a test temperature of –40°C is at the lower limit of requirements (0.15 mm).

Thus, the desired combination of strength, ductility, toughness, and performance characteristics at low temperatures in rolled sheets made of low-alloy, shipbuilding steels with a thickness of up to 50 mm with a guaranteed yield point of 355–390 MPa is ensured by the formation of a ferrite–bainite structure that is quasi-uniform across the thickness of the rolled metal sheet, which has structural components that are similar in morphology; the developed subgrain structure (confirmed by the presence of a significant number of small-angle boundaries (around 33–35%)); the fraction of quasi-polygonal ferrite below 45%; and the fraction of bainite with the lath morphology below 15%. According to the EBSD data, the number of structural components with sizes of no more than 5 and 10 μm at a given angle of tolerance θ_t = 5° should be about 30 and 70%, respectively; moreover, the mean size of structural components over the entire thickness of the rolled-metal sheet should be 6–8 μm.

Low-alloy steels of strength categories of 420–460 MPa. Rolled sheets made of low-alloy, cold-resistant steels with a thickness of up to 50–60 mm and a guaranteed yield point of 420–460 MPa are also manufactured with the TMP + RC technology to temperatures that correspond to the middle and lower parts of the bainitic transformation. At the final (finishing) stage of rolling, an ascending schedule is used in a narrow (no more than 30°C) temperature range that lies 150–200°C below the threshold temperature of austenite recrystallization, which ensures the formation of a developed subgrain structure in austenite due to polygonization and fragmentation processes.

An increase in the strength characteristics with respect to the properties of steels that belong to strength categories of 355–390 MPa is achieved via a decrease in the fraction of quasi-polygonal ferrite (no more than 30%) in the structure of sheet products with a guaranteed yield point of 420–460 MPa and the formation of a predominantly bainitic structure with a granular morphology (Figs. 11a and 11c), as well as a slight increase in the amount of lath bainite from 15 to 20% over the entire thickness of the sheet. In this case, the mean size of structural components at a given tolerance angle of 15° is 13.3 μm, and the size of structural components at a given tolerance angle of 5° barely changed as compared to the corresponding parameter (7.8 μm) in steels of strength categories of 355–390 MPa (Fig. 11b).

The formation of this kind of structure without a significant number of large regions of lath bainite formed within the former unrecrystallized austenite grains elongated along the rolling direction gives thick rolled-sheet products a high cold resistance. The values of the impact work measured up to a test temperature of –80°C are not less than 200 J. The obtained values of the critical brittleness temperatures T_{cb} = –20°C and ZDT = –75°C fully comply with the requirements for steels with the Arc40 index. At the same time, a high level of crack resistance is achieved: the mean displacement of the CTOD at a test temperature of –40°C is 0.72 mm as compared to the necessary value of at least 0.20–0.25 mm according to the requirements.

Upon the formation of an anisotropic ferrite–bainite structure with a significant fraction of extended regions of lath bainite (Fig. 11f) with ferrite grains (Fig. 11d), the ductile-to-brittle transition temperature T_{cb} rises to +15°C, the zero-plasticity temperature (ZDT) rises to –25°C, and the mean displacement of CTOD at a test temperature of –40°C

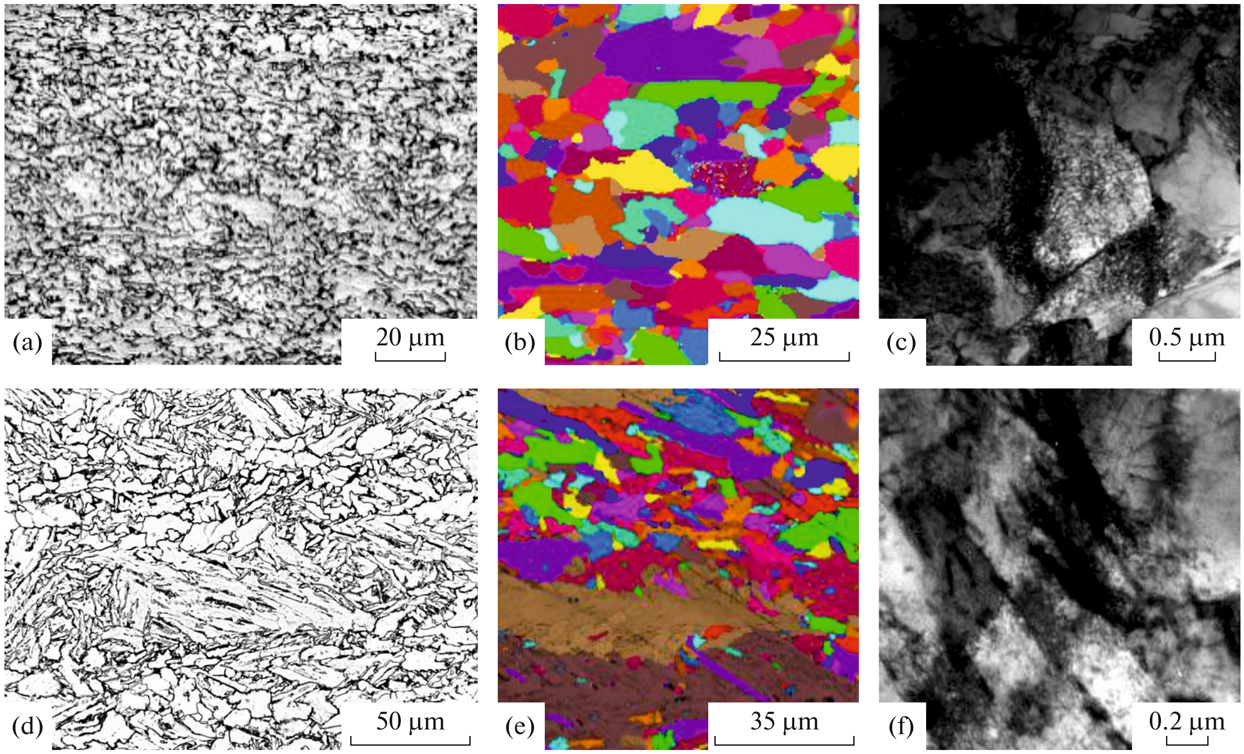

Fig. 11. Structure of a rolled sheet made of low-alloy steel belonging to a strength category of 420–460 MPa manufactured with the TMP + RC technology (a–c) in the optimal regimes and (d–f) with deviations from the optimal regimes.

decreases to 0.08 mm. These values do not meet the requirements.

Thus, the combination of the required strength, ductility, toughness, and performance characteristics at low temperatures in rolled sheets with a thickness of up to 50–60 mm made of low-alloy steels with a guaranteed yield point of 420–460 MPa is achieved by the formation of a ferrite–bainite structure that is quasi-uniform across the thickness of the rolled-metal sheet with bainite of predominantly granular morphology and a quasi-polygonal ferrite content of up to 20–30%, which is characterized by a developed substructure of deformation origin (the share of SAB is about 40%). The fraction of bainite with the lath morphology should not exceed 20%.

An increase in the multiplicity of deformation during the rolling of ferritic–bainitic steels leads to refinement of structural components and an increase in the quantity of finely dispersed, quasi-polygonal ferrite, which contributes to an increase in the visco-plastic characteristics. When a predominantly bainitic structure is formed, an increase in the deformation fractionalization does not have a positive effect. In light of the above, the maximum possible compression in one pass should be achieved during rolling in the production of bainitic steels [75].

The formation of carbides and carbonitrides with a particle size of 10–50 nm in austenite and in the final structure of microalloyed steel is an additional method of structure refinement. The study of the effect of vanadium on the morphology of the bainitic structure showed that the additional introduction of vanadium into steel in an amount of 0.03 wt % increases the number of nucleation sites for the formation of a new phase (vanadium nitrides and carbonitrides) before the $\gamma \rightarrow \alpha$ transition, enhances the process of quasi-polygonal ferrite precipitation, increases the critical point Ar_1, and ensures the formation of predominantly granular structures [95].

Moreover, the effect of precipitation hardening is observed in low-alloy steels with a bainitic structure of the granular type and a developed substructure (after thermomechanical processing with stepwise cooling) during subsequent tempering, and the introduction of 0.03 wt % vanadium increases the strength by 40–50 MPa. Technologies based on the principles of controlled precipitation of carbides and carbonitrides during stepwise cooling or high-temperature tempering are very promising for the development of the production of high-strength, low-alloy steels (which was used in the development of the QHR + T technology for the production of rolled-metal sheets with a thickness of 61–100 mm thick from steels with a guaranteed yield strength of 420–460 MPa) [96].

The development of temperature–deformation rolling schemes for the manufacture of thick sheets (over 60 mm) requires a special approach.

With the use of one-step deformation of rolled sheets with a thickness of 80–100 mm from low-alloy steels manufactured with the ZPN + O technology, a tempered bainite–martensite structure with the size of lath bainite regions up to 80 μm and a bainite structure in the form of bainite regions of granular and lath morphology with large lath bainite regions up to 120 μm in size and ferrite grains up to 20 μm in size form in the surface layers and in the middle part across the sheet thickness, respectively. Such a structure does not provide stable performance parameters for the impact work at a test temperature of –60°C in samples cut from the middle part of the thickness of a rolled sheet (14–213 J with a mean value of 146 J) [96].

Deformation with a two-step scheme makes it possible to obtain a bainite structure in rolled-metal sheets with an acceptable degree of heterogeneity over its thickness, as well as tempered, lath-type, bainite regions with dimensions of no more than 30–40 μm in sheets with a thickness of 80 mm and 50–60 μm in sheets with a thickness of 100 mm. The anisotropy coefficient[6] of such a structure over the thickness is K_a = 0.78–1.09 in rolled products with a thickness of 80 mm and K_a = 0.87–1.14 in rolled products with a thickness of 100 mm; the corresponding levels of the impact work at a test temperature of –60°C in the middle part across the sheet thickness are 189–207 and 168–198 J, respectively [96].

Economically alloyed steels of strength categories of 500–620 MPa. A high level of strength and viscoplastic properties (relative elongation up to 22.5%, impact work not less than 200 J at a test temperature of –60°C) of rolled-metal sheets with a thickness of up to 50 mm from economically alloyed cold-resistant steels with a guaranteed yield point of 500–620 MPa is obtained with the use of special QHR + T regimes and, less often (in limited thickness ranges), after TMP + RC.

Given the higher content of alloying elements in chromium–nickel–copper–molybdenum steels microalloyed with niobium, the main mechanism for the refinement of the austenitic structure at the rough (high-temperature) stage of rolling is multiple static recrystallization, which can be fully ensured by a gradual increase in the deformation pauses between successive compressions with a temperature decrease in the region of partial dynamic and static recrystallization of austenite (1150–1000°C) at the rough stage [55]. However, it is extremely important to exclude excessively long holding times between passes and to limit the duration of intermediate cooling to prevent the process of cumulative recrystallization in austenite.

In the finishing (final) stage of rolling of steels of strength categories of 500–620 MPa with the Arc40 index, it is advisable to use a descending temperature schedule of the rolling regime in a sufficiently narrow temperature range (no more than 40–50°C) that starts at a temperature 50–70°C lower than the threshold temperature of austenite crystallization; this excludes strain hardening of austenite and ensures the creation of a developed subgrain structure in austenite due to the processes of dynamic polygonization and fragmentation.

The formation of a bainite–martensitic structure that is quasi-isotropic over the thickness of rolled sheets makes it possible to ensure the high strength of the steel, together with the required levels of crack resistance and cold resistance. Moreover, the structure represents a mixture of bainite of granular and lath morphology with a mean size of structural components of 12.2 μm at a given tolerance angle of 15° and 9.4 μm at a given tolerance angle of 5° (Fig. 12b) and martensites of the lath and nonlath types, the proportion of which depends on the thickness of the sheet metal. The impact-work values at a test temperature of –80°C are in the range of 159–288 J. The obtained values of the ductile-to-brittle transition (T_{cb} = –55°C) and zero plasticity (ZDT = –75°C) temperatures meet the requirements for cold-resistant steels with the Arc40–Arc50 indices. The mean displacement of CTOD at a test temperature of –40°C is 0.45 mm, as compared to the necessary value of at least 0.25 mm according to the requirements.

There is a significant deterioration in the cold resistance of steel—in particular, an increase in the T_{cb} temperature to +20°C and ZDT to –40°C—if the bainite component in the structure is extremely heterogeneous due to the following nonuniform processes:

— recrystallization in austenite during hot-plastic deformation, as a result of which large regions of lath bainite form within the former unrecrystallized austenite grains during subsequent quenching cooling;

— softening during tempering, which leads to the formation of large tempered regions adjacent to regions with a preserved lath structure (Figs. 12d and 12f).

In this case, EBSD analysis shows an increase in the mean size of structural components to 19.2 μm with misorientations of more than 15° at the boundaries (Fig. 12e).

Thus, the required combination of the strength, ductility, toughness, and performance characteristics at low temperatures in rolled sheets with a thickness of up to 50 mm from economically alloyed steels with a guaranteed yield point of 500–620 MPa is ensured by the formation of a bainite–martensite structure that is quasi-isotropic across the thickness of the rolled sheet with a high fraction of small-angle boundaries at a level of about 45–55% and the presence of bainite of lath and granular morphology with a martensite fraction in the range from 10 to 50% in the absence of a ferrite component. For a given tolerance angle of θ_t =

[6] The anisotropy coefficient characterizes the presence and intensity of the preferable orientation of structural components [97–99].

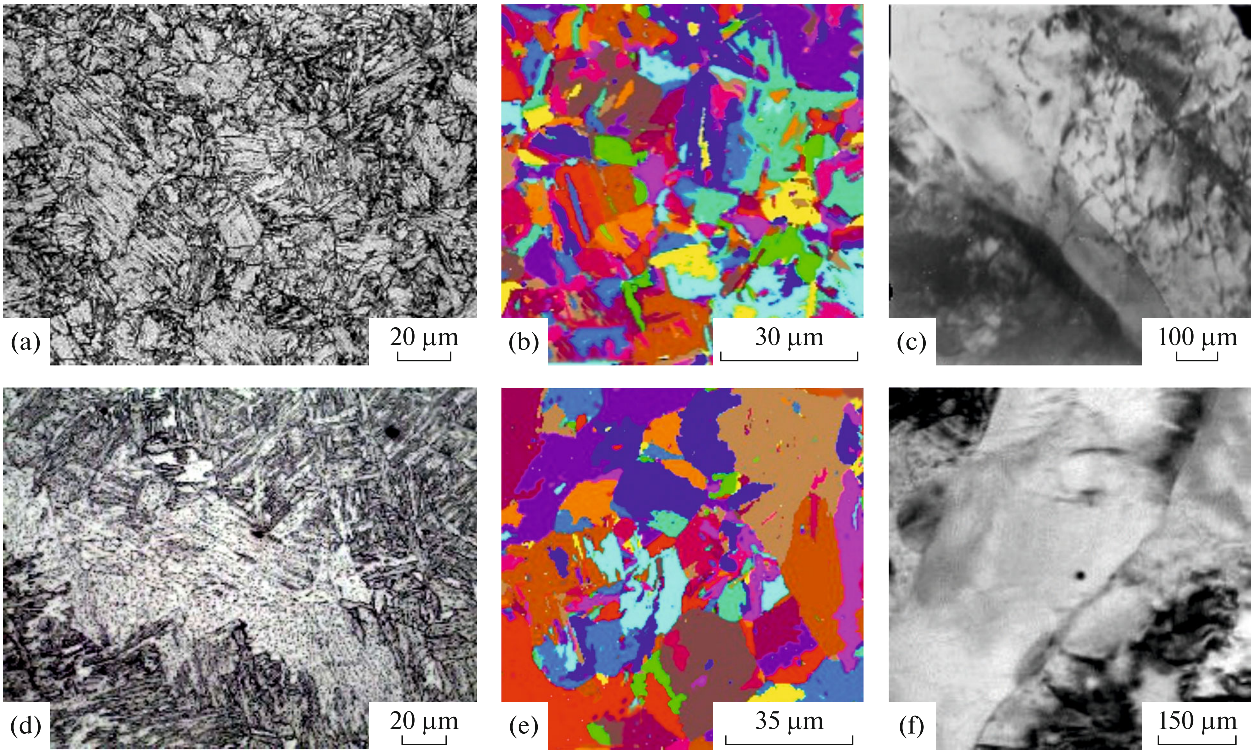

Fig. 12. Structure of a rolled sheet made of economically alloyed steel belonging to a strength category of 500–620 MPa manufactured with the QHR + T technology (a–c) in the optimal regimes and (d–f) with deviations from the optimal regimes.

5°, the number of structural components with a size of no more than 5 μm should be about 20–25%, and the number of structural components with a size of no more than 10 μm should be at least 55%; the mean size of structural component should be 7.5–11 μm throughout the entire thickness of the rolled metal sheet.

Economically alloyed steels of strength categories of 690–960 MPa. Rolled sheets with a thickness of up to 50 mm from economically alloyed cold-resistant steels with a guaranteed yield point of 690–960 MPa are produced with the technology of QHR + T.

As established in [72], the temperature of the onset of dynamic recrystallization in high-strength steels of a chromium–nickel–copper–molybdenum alloying composition microalloyed with niobium is above 1100°C at strain degrees of at least 20% per pass. Therefore, it is impossible to ensure the completeness of the recrystallization process under the existing technological limitations of rolling equipment. This leads to undesirable structural inhomogeneity, which is characterized by coexistence of small, recrystallized grains with large, deformed ones, and the difference in grain size is not eliminated, even during metadynamic recrystallization. In steels of these strength categories, as well as in weaker steels with a guaranteed yield strength of 500–620 MPa, static recrystallization remains the only mechanism for the formation of a homogeneous, fine-grained, austenite structure.

In addition to the refinement of the initial austenite grain at the high-temperature (rough) stage of hot rolling due to the controlled static recrystallization processes in the interdeformation pauses, it is important to select temperature–deformation rolling schemes at the low-temperature (finishing) stage of rolling.

The principle proposed in [73] to control the structural parameters in low-alloy steels via variation of the temperature scheme of deformation at the finishing stage with ascending, descending, and isothermal schedules was also applied to high-strength steels. As shown in [74], the procedure of hot-plastic deformation at the final (finishing) stage of rolling should be carried out with a constant degree of compression of 12–13% per pass at a constant temperature that is 50–70°C lower than the threshold temperature of static recrystallization in order to form austenite grains with a developed substructure of deformation origin with a size of no more than 16 μm. The use of such a technological procedure makes it possible to increase the strength characteristics of steel by 50–100 MPa without changes in the alloying level.

Hardened and tempered steel is used in the most critical components of marine structures. The required level of mechanical characteristics in a steel with a bainite–martensite structure can be achieved only via high-temperature tempering. Low-temperature tempering provides high strength characteristics and low values of ductility and toughness.

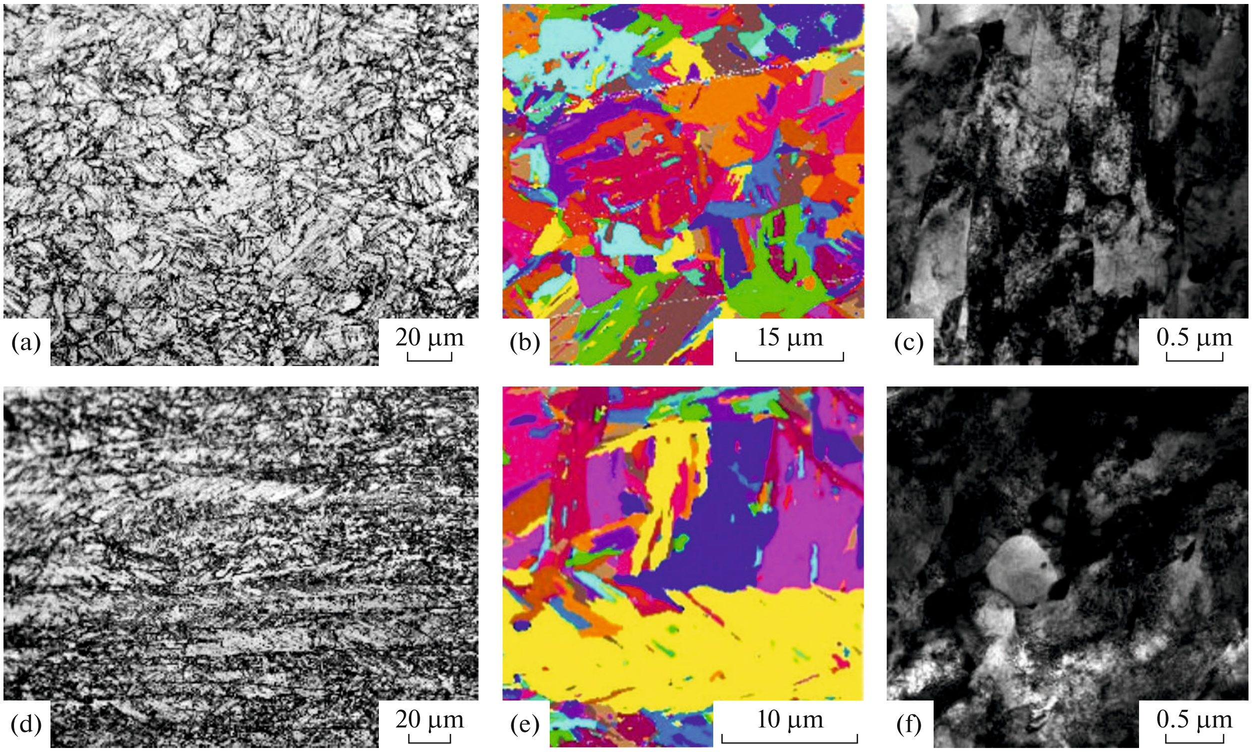

Fig. 13. Structure of a rolled sheet made of economically alloyed steel belonging to a strength category of 690–750 MPa manufactured with the QHR + T technology (a–c) in the optimal regimes and (d–f) with deviations from the optimal regimes.

The formation of a dispersed bainite–martensite structure (Fig. 13a) predominantly composed of lath martensite (LM) (up to 75–80%) with a high dislocation density (2.3–2.9) × 10^{14} m^{-2} (Fig. 13c) provides both the required strength characteristics and the required cold- and crack-resistance characteristics. The second structural component in such steels is self-tempered martensite (high-temperature martensite), which has a dislocation density close to that observed in LM, which contains dispersed carbides 40–50 nm in size, and a bulk density of 1.5 × 10^{20} m^{-3}. Such a structure makes it possible to obtain a guaranteed yield strength of 690 MPa and higher, a relative elongation of at least 15%, and an impact work of at least 80 J in the temperature range of –40 to –60°C. The obtained values of critical brittleness temperatures (T_{cb} = –55°C and ZDT = –75°C) are close to the values for weaker steels that belong to strength categories of 500–620 MPa and meet the requirements for cold-resistant steels with the Arc40–Arc50 indices.

In economically alloyed steels with a guaranteed yield point of 690–960 MPa with a bainite–martensite structure, the predominantly shear-induced nature of transformation contributes to a significant refinement of structural components with boundaries both with misorientation angles of more than 15° and more than 5° (packets, blocks, and subblocks) [100–102]. This makes it possible to maintain the required fracture toughness (the CTOD is no less than 0.30 mm) at a given strength level. Thus, the mean sizes of structural components in steels of strength categories 690–750 (Fig. 13b) and 890–960 MPa are 9.7 and 9.0 μm, and 7.0 and 5.1 μm at given tolerance angles of 15° and 5°, respectively.

Upon technological deviations from the optimal temperature–deformation regimes of rolling in economically alloyed steels with a guaranteed yield point of 690–750 MPa, an inhomogeneous, predominantly bainite structure with extended regions of bainite with lath morphology can be formed (Figs. 13d and 13e) in the presence of a ferrite component (Fig. 13f). This leads to a decrease in all mechanical characteristics with a simultaneous increase in the values of the critical brittleness temperatures T_{cb} to 0°C and ZDT to –45°C.

It should be noted that the carbon equivalent, which indirectly determines the weldability and the level of alloying, is substantially limited for steels with a guaranteed yield strength of 890–960 MPa. Research on the creation of economically alloyed, shipbuilding steels for civil purposes with a yield point of 890 MPa or higher is currently being carried out in Russia and abroad [103].

Thus, the required combination of the strength, ductility, toughness, and performance characteristics at low temperatures in rolled-sheet products up to 50 mm in thickness from economically alloyed steels with

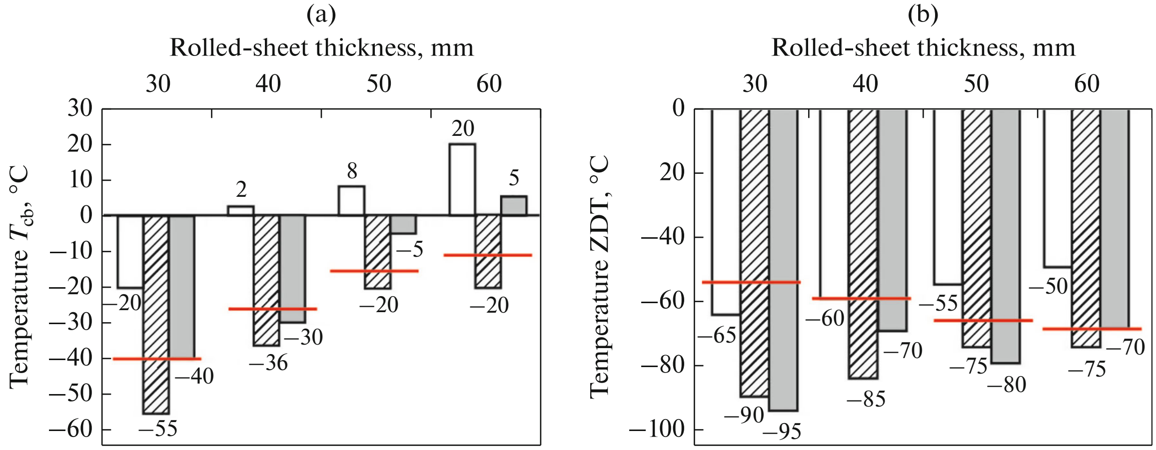

Fig. 14. Minimum values of the critical brittleness temperatures (a) T_{cb} and (b) ZDT for rolled sheets with a thickness of 30–60 mm from low-alloy steels with different contents of the main alloying elements (the lines correspond to the specification requirements): unfilled columns correspond to (1.35–1.40) wt % Mn without Ni and Cu; shaded columns correspond to (1.15–1.20) wt % Mn + (0.85–0.90) wt % (Ni + Cu); gray columns correspond to (1.15–1.20) wt % Mn + (1.15–1.20) wt % (Ni + Cu) [104].

a guaranteed yield point of 690–960 MPa is ensured by the formation of a bainite–martensite structure that is quasi-isotropic across the thickness of the rolled-metal sheet with a high dislocation density and a developed substructure with a lath martensite fraction of up to 75–80%.

Moreover, the parameters of the bainite–martensite structure for economically alloyed steels of strength categories of 690–960 MPa are determined by their alloying composition.

ESTABLISHED RELATIONSHIPS BETWEEN ALLOYING, VARIOUS STRUCTURAL PARAMETERS, AND COLD- AND CRACK-RESISTANCE CHARACTERISTICS AT LOW TEMPERATURES

The performance characteristics required for high-strength Arc steels at low temperatures are provided in thick rolled-sheet products only in the case when all of the described developed technological methods of rolling and quenching from hot rolling are fulfilled.

Alloying–Performance Characteristics Relationships

It was established that the ranges of the contents of carbon and the main alloying elements specified in GOST 52927–2015 and Rules of the RMRS (State standards of Russia) and the choice of rational microalloying in rolled-sheet products made of cold-resistant, low-alloy (up to 100 mm in thickness) and economically alloyed (up to 50 mm in thickness) steels, including those with the Arc index, should be narrowed to ensure a stable combination of strength, impact work across the thickness, and performance characteristics determined at low temperatures on full-thickness samples [104].

Thus, the specified cold-resistance characteristics, i.e., the ductile-to-brittle transition temperatures T_{cb} and the zero plasticity temperature ZDT, as well as high crack-resistance characteristics according to the CTOD criterion at temperatures ranging from –40 to –60°C, are achieved in the entire range of considered sheet thicknesses of 30–60 mm in low-alloy steels alloyed with Mn with a guaranteed yield point of 355–460 MPa when the total (Ni + Cu) content is not higher than 1.0 wt% [104] (Fig. 14).

It is much more difficult to maintain the required set of mechanical properties in combination with the required cold-and crack-resistance characteristics in economically alloyed steels that belong to strength categories of 500–960 MPa, because the selected ranges of the content of the main alloying elements of high-strength steel (manganese, nickel, copper, chromium, and molybdenum) should provide the required hardenability across the entire rolled sheet thickness up to 50 mm (and up to 100 mm in accordance with the latest requirements of Rules of the RMRS) upon limitation of the carbon equivalent value C_{eq}.

For example, in economically alloyed steels with C_{eq} values of 0.41% (about 1 wt % (Ni + Cu + Mo)) and 0.46% (0.5 wt % Cr + 2.0 wt % (Ni + Cu + Mo)), it is possible to ensure a level of strength characteristics that correspond only to the strength category of 500 MPa, despite the high level of impact work at a test temperature of –60°C. An increase in the C_{eq} values to 0.51–0.61% makes it possible to provide a strength level that meets the requirements for steels of strength categories of 620–750 MPa. In this case, high values of the impact work at test temperatures down to –80°C can be achieved with various alloying compo-

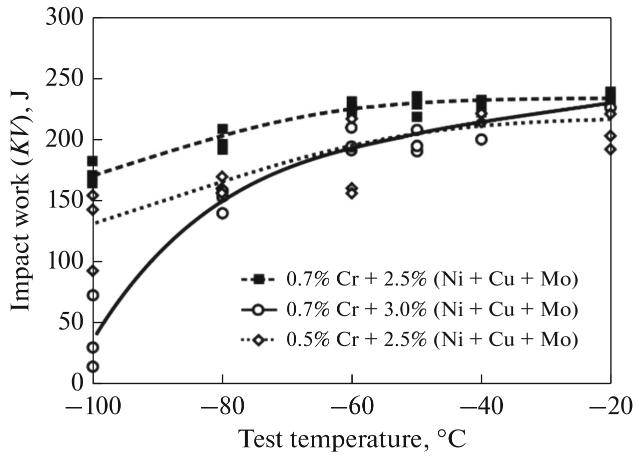

Fig. 15. Effect of the Cr content and the total (Ni + Cu + Mo) content on the impact-work value at low temperatures [104].

sitions: 0.5 wt % Cr + 2.5 wt % (Ni + Cu + Mo) with C_{eq} = 0.51%; 0.7 wt % Cr + 2.5 wt % (Ni + Cu + Mo) with C_{eq} = 0.54%; and 0.7 wt % Cr + 3.0 wt % (Ni + Cu + Mo) with C_{eq} = 0.59% (Fig. 15) [104]. At the same time, low values of the ductile-to-brittle transition temperature (T_{cb} = –55...–72°C) and the zero plasticity temperature (ZDT = –75...–105°C) are achieved in rolled sheets with a thickness of 40–50 mm, which meet the requirements for cold-resistant steels with the Arc40–Arc60 indices.

The cooling rate in the middle part across the thickness in rolled-metal sheets made of economically alloyed steels with a guaranteed yield point of 500–750 MPa decreases with an increase in the thickness to above 50 mm. This increases the likelihood of the formation of a structure that is nonuniform across the thickness and contains large regions of granular and lath bainite due to an insufficient content of alloying elements. In this regard, it is advisable to increase the Cr content to 1.10–1.20 wt % with a total Ni + Cu content of 2.25–2.35% in order to increase the hardenability of steel during the quenching of the rolled-metal sheet within a thickness range of 51–100 mm [104].

It should be noted that the use of economical alloying in the manufacture of thick rolled-sheet products from high-strength steels significantly limits the temperature range in which the required CTOD values are provided.

Relationship of Structural Parameters and Performance Characteristics

In addition to the general concepts of the optimal structure already described above, other important requirements that are much more stringent than those previously imposed on shipbuilding steels of the D, E, and F cold resistance categories must be met to guarantee the performance characteristics of rolled-metal sheets with a thickness of up to 100 mm.

In addition to the development of quantitative requirements for the key structural parameters, the standardization of their range across the rolled-sheet thickness to characterize the permissible degrees of inhomogeneity and anisotropy (the difference between the minimum and maximum values) is another important approach to the assessment of the quasi-isotropic structure of cold-resistant steels. It was not taken into account earlier, but it determines the achievement of high cold- and crack-resistance characteristics. At the same time, the key structural parameters that determine the performance characteristics differ significantly for steels with different structures.

Low-alloy steels with a ferrite-bainite structure. To ensure the performance characteristics in rolled-metal sheets made of low-alloy steels with a yield point of at least 355–460 MPa manufactured with the technology of thermomechanical processing with rapid cooling, complex requirements must be met for various morphological and crystallogeometric parameters of the structure, e.g., the structural anisotropy, morphology, dispersity and ratio of structural components, the volume fraction and size of bainite regions with the lath morphology, and a number of fine structural parameters assessed via transmission electron microscopy (TEM) and an automated analysis of EBSD analysis [105, 106].

Figure 16 illustrates the effect of structural anisotropy determined by the anisotropy coefficient K_a [97–99] on the results of impact-bending tests and the values of the critical temperatures T_{cb} and ZDT for thick rolled sheets made of low-alloy steels. Thus, the K_a values in rolled-metal sheets with a thickness of up to 100 mm made of low-alloy steels with a yield point of at least 355–460 MPa should not exceed 1.35 (Fig. 16a) with a maximum allowable difference of 0.50 in the sheet thickness to ensure a high impact work at test temperatures of –60...–80°C [104, 105].

At the same time, it is necessary to tighten the requirements for the permissible value of K_a to no more than 1 (Fig. 16b) to obtain guaranteed performance characteristics that meet the requirements for cold-resistant steels with the Arc40 index [105]. It is important that these requirements are included in the 2019 edition of Rules of the RMRS [23, 24].

Figure 17 illustrates the effect of the fine parameters of the ferrite–bainite structure in a rolled-metal sheet from low-alloy steel with a thickness of 50–60 mm, which are determined at various points across the thickness of the sheets with EBSD analysis, on the T_{cb} and ZDT temperature values and the CTOD at a test temperature no higher than –40°C.

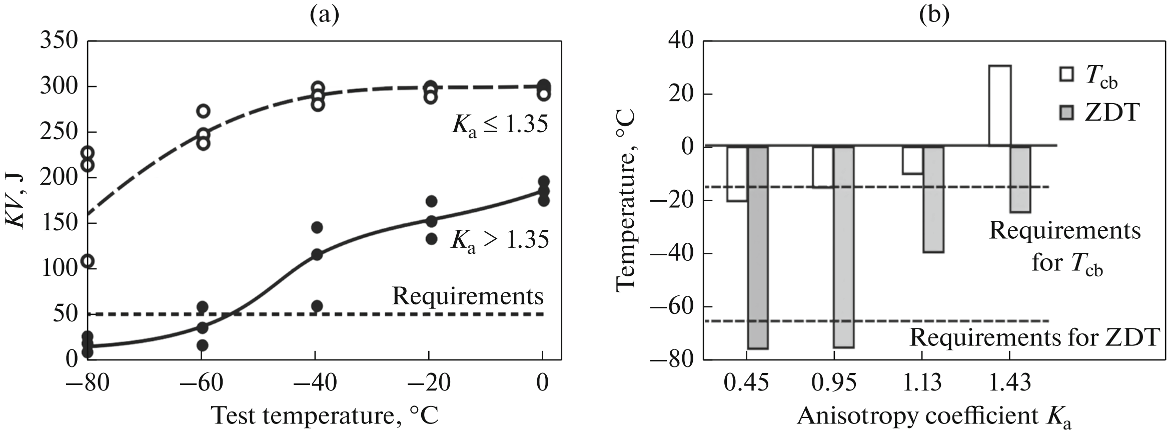

Fig. 16. Effect of the coefficient of structural anisotropy of low-alloy steel on the subsequent cold-resistance characteristics of rolled-metal sheets: (a) impact work at low test temperatures and (b) critical brittleness temperatures T_{cb} and ZDT [105].

It was established in [104] that the guaranteed performance characteristics of rolled sheets made of low-alloy steels of strength categories of 355–460 MPa with the Arc index manufactured with the TMP + RC technology are achieved with the following restrictions on the structural parameter values and their maximum permissible difference (Δ) across the sheet thickness:

— a fraction of small-angle boundaries of 25–40% SAB for Δ% SAB $\leq$ 15% (Fig. 17a);

— a fraction of large-angle boundaries that corresponds to misorientation angles of $\theta = 50°–62.5°$ of no less than 20% LAB for Δ% LAB $\leq$ 15% (Fig. 17b);

— a mean size D_m of structural components at a given tolerance angle of $\theta_t = 5°$ of no more than 11 μm for $\Delta D_m \leq 5$ μm (Fig. 17c);

— a maximum size D_{max} of structural components at a given tolerance angle of $\theta_t = 5°$ of no more than 20 μm for $\Delta D_{max} \leq 5$ μm (Fig. 17d);

— a fraction of structural components with a D size of no more than 5 and 10 μm at a given tolerance angle of $\theta_t = 5°$ of no less than 20 and 65%, respectively (not more than 25% for $\Delta\% \ D_{\leq 5\ \mu m}$ and $\Delta\% \ D_{\leq 10\ \mu m}$) (Figs. 17e and 17f);

— a mean curvature GAM_m of the crystal of no more than 0.65° for $\Delta GAM_m \leq 0.25°$ (Fig. 17g).

Economically alloyed steels with bainite and bainite–martensite structures. The main structural parameters for high-strength, economically alloyed steels with a yield point of at least 500–960 MPa produced with QHR + T are the mean size $D_{FAG\ m}$ of the former austenite grain (FAG) and its heterogeneity degree $D_{FAG\ max}–D_{FAG\ min}$ across the thickness of the rolled-metal sheet [74], the mean size and fraction of structural components with a given size at tolerance angles of $\theta_t = 5°$ and 15°, and the ratio of structural components (martensite and bainite of various morphologies with estimation of the dislocation density, and the volume fraction of carbide precipitates in the body and along grain and subgrain boundaries).

As shown in [106–109], it is important to create effective grains separated by large-angle boundaries during the formation of a bainitic or bainitic-martensitic structure. The obtainment of high cold resistance and crack resistance to high-strength steels depends on such crystalline components, which can be not only packets but also individual blocks and structural components with boundaries of deformation origin and misorientation angles θ of at least 5° between them.

However, this requires a more detailed study of the peculiarities of the internal structure of bainite and martensite in chromium–nickel–copper–molybdenum steels (laths, packets, and blocks [100, 109–111]), which became possible with the development and application of new methods for the certification of the structure of high-strength steels with EBSD analysis [101, 109, 112–114] in combination with TEM data.

COLD RESISTANCE CHARACTERISTICS OF ROLLED SHEETS FROM LOW-ALLOY AND ECONOMICALLY ALLOYED STEELS

Figures 18 and 19 show the cold-resistance characteristics (impact work values at a test temperature of –60°C, critical ductile-to-brittle transition temperature T_{cb}, and zero plasticity temperature ZDT) for rolled-metal sheets with thickness of 40–50 mm from Arc steels with a guaranteed yield strength of 355 to 750 MPa manufactured with the optimized TMP + RC and QHR + T technological processes [96].

As shown in [96], a high level of impact work at a test temperature of –60°C can also be achieved in rolled sheets with a thickness of 80–100 mm from low-alloy steels of the F cold-resistance category with guaranteed yield points of 355–390 MPa (according

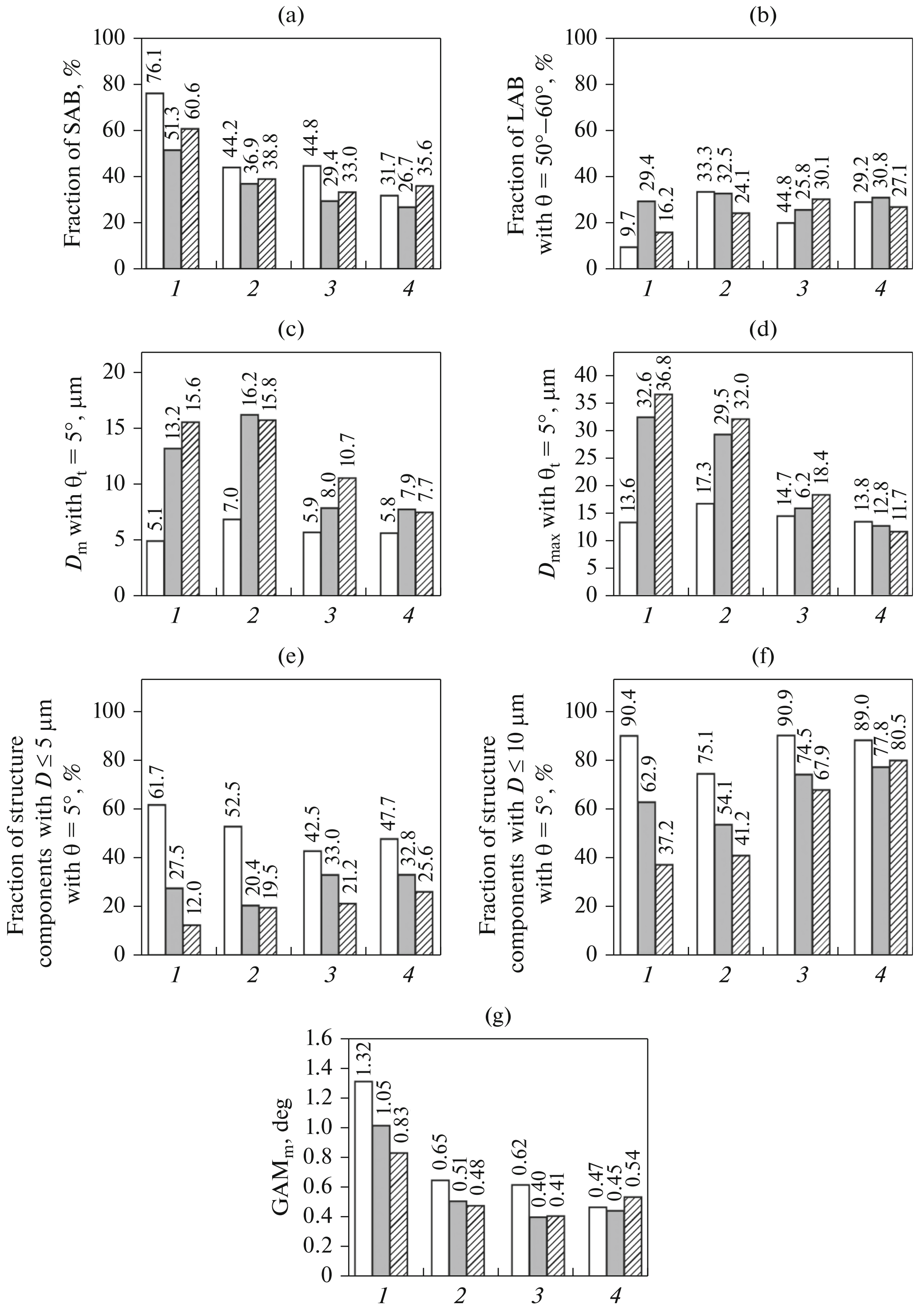

Fig. 17. Effect of the parameters of the ferrite–bainite structure of rolled-metal sheets with a thickness of 50–60 mm from low-alloy steel as determined from the EBSD analysis on the performance characteristics (T_{cb}, ZDT, and CTOD): (*1*) T_{cb} = + 15°C, ZDT = –75°C, and CTOD^{-40} = 0.13 mm; (*2*) T_{cb} = +10°C, ZDT = –50°C, and CTOD^{-40} = 1.24 mm; (*3*) T_{cb} = –15°C, ZDT = –65°C, and CTOD^{-40} = 0.23 mm; (*4*) T_{cb} = –20°C, ZDT = –75°C, and CTOD^{-40} = 1.2 mm. Unfilled columns show the parameters near the surface, gray columns show the parameters in a quarter of the sheet thickness, and shaded columns show the parameters in the middle part across the sheet thickness [104].

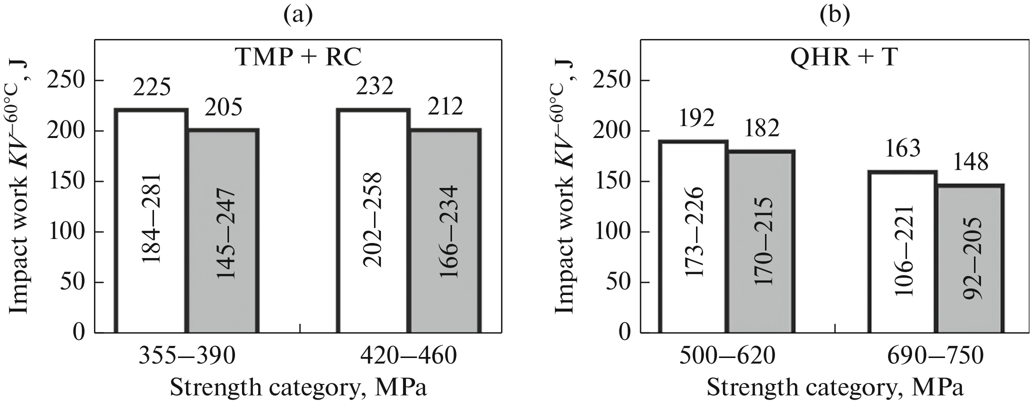

Fig. 18. Impact-work values at test temperatures of –60°C for rolled sheets with a thickness of 40–50 mm from low-alloy and economically alloyed steels with the Arc index manufactured with (a) TMP + RC and (b) QHR + T technologies. Unfilled columns show the parameters near the surface and gray columns show the parameters in the middle part across the sheet thickness [96].

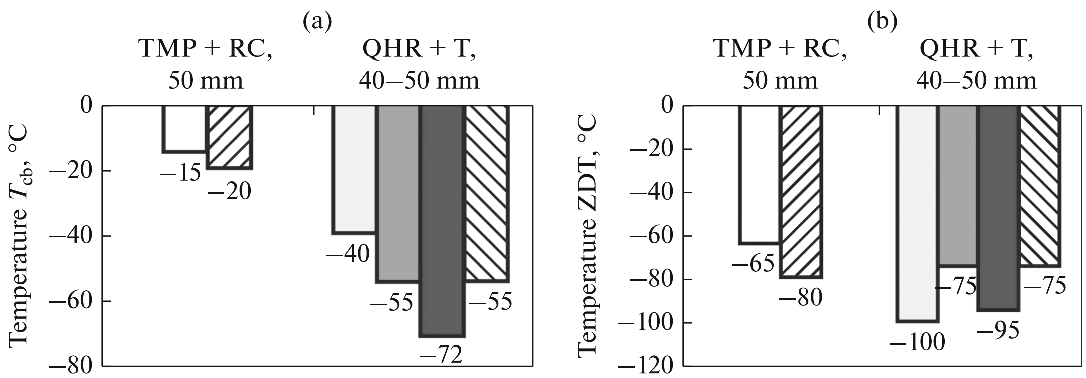

Fig. 19. Results of the determination of the critical brittleness temperatures (a) T_{cb} and (b) ZDT for rolled sheets with a thickness of 40–50 mm from low-alloy and economically alloyed steels with the Arc index of the following strength categories: 355–390 MPa (unfilled columns); 420–460 MPa (left hatched columns); 500 MPa (light-gray columns); 620 MPa (dark-gray columns); 690 MPa (dark columns); 750 MPa (right hatched columns) [96].

to the TMP + RC technology) and 420–460 MPa (according to the QHR + T technology) (Fig. 20).

At the same time, the obtained cold-resistance characteristics of rolled sheets made of low-alloy and economically alloyed steels of various strengths are guaranteed by the developed energy-saving technologies (TMP + RC and QHR + T) due to the formation of a specified quasi-isotropic structure with admissible degrees of inhomogeneity and anisotropy across the sheet thicknesses.

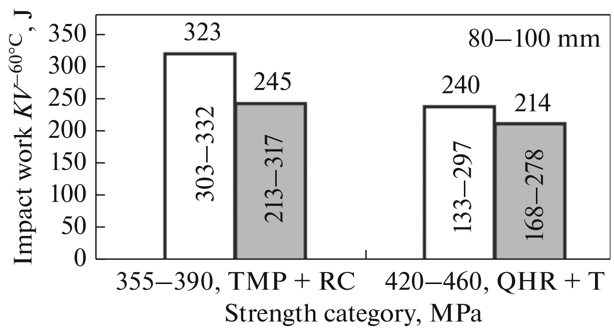

Fig. 20. Impact-work values at test temperatures of –60°C for rolled-metal sheets with a thickness of 80–100 mm from low-alloy steel; unfilled columns show the values near the surface and gray columns show the values in the middle part across the sheet thickness [96].

COLD- AND CRACK-RESISTANCE OF HEAT-AFFECTED ZONES OF WELDED JOINTS

With a high level of mechanical properties and performance characteristics of the BM, the obtainment of stable cold- and crack-resistance parameters for the

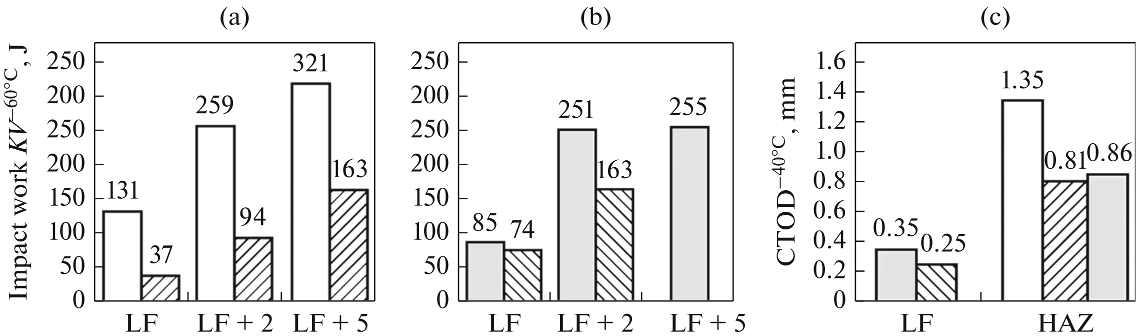

Fig. 21. Mean values of (a, b) of the impact work *KV* at a test temperature of –60°C and (c) the CTOD in the HAZs of welded joints from low-alloy and economically alloyed Arc steels of the following strength categories: 355–390 MPa with a heat input of 3.5 kJ/mm (unfilled columns); 355–390 MPa with a heat input of 6 kJ/mm (left hatched columns); 500 MPa with a heat input of 3.5 kJ/mm (light-gray columns); 750 MPa with a heat input of 2.6 kJ/mm (right hatched columns) [96].

metal of the HAZ of welded joints remains a challenge.

The performed set of tests of welded joints confirmed the satisfactory weldability of rolled-metal sheets made of cold-resistant steels with a guaranteed yield strength of 355–750 MPa at a heat input of up to 3.5 kJ/mm. For welded joints of rolled-steel sheets with a thickness of 40–60 mm from low-alloy steels of strength categories 355–390 MPa (Fig. 21a) and economically alloyed steels of strength categories of 500–750 MPa (Fig. 21b), cold-resistant steels provide a high level of impact work at a test temperature of –60°C on samples cut both along the line of fusion (LF) and at a distance of 2 and 5 mm from it (LF + 2 and LF + 5) [96].

The obtained mean values of the CTOD of welded joints are indicative of high crack resistance of the HAZ metal at a test temperature of –40°C for both low-alloy and economically alloyed cold-resistant steels (Fig. 21c) [96].

An increase in the heat input during welding to 6 kJ/mm during the production of welded joints from low-alloy steels leads to a significant decrease in the mean impact-work values. The lowest values were obtained for samples with a cut along the LF due to the formation of a coarse-grained bainite structure up to 200 μm in size inside the former austenite grains with an intense precipitation of the carbide phase along their boundaries. At the same time, the impact work and crack resistance of the HAZ metal substantially exceeds the requirements from Rules of the RMRS [96].

The study of the structure and properties of HAZs of welded joints made of low-alloy and economically alloyed cold-resistant steels for applications in the Arctic region is discussed in more detail in [115–125].

The high cold resistance and crack resistance of the HAZ metal indicates that the developed materials can be used for welded structures operating in the difficult climatic conditions of the Arctic region. The use of high-performance welding processes without a loss of cold resistance of the metal in the HAZ of welded joints makes it possible to reduce the labor intensity of welding in shipbuilding plants by at least 30%.

INDUSTRIAL IMPLEMENTATION

The production of rolled sheets from cold-resistant steels with a wide range of strength characteristics was established at the Public Joint Stock Company (PAO) Magnitogorskii Metallurgicheskii Combinat, PAO Severstal', the Limited Liability (OOO) OMZ-Spetsstal', and AO Ural'skaya Stal' (Table 6). The developed materials have already been used to build an icebreaker fleet and marine and engineering equipment operating in the Arctic conditions (including ice navigation vessels, ice-resistant offshore drilling platforms, and lifting and transporting equipment that allow the exploration and development of oil and gas fields and coastline territories). In 2010–2018, about 500 000 t of high-quality sheet metal from low alloyed and economically alloyed steels of cold-resistant grades belonging to strength categories from 315 to 690 MPa were supplied following a request from the leading shipbuilding enterprises of the Russian Federation [96].

Cold-resistant, rolled sheets were used to build the world's largest nuclear icebreakers of the Arktika, Siberia, and Ural projects; a diesel-electric icebreaker of the Aker ARC 130 A project; the world's largest multifunctional linear diesel-electric icebreaker, the Viktor Chernomyrdin; and other vessels. They were also recommended for use in the construction of the Leader icebreaker, which has a nuclear reactor capacity of 110 MW.

New, highly reliable, cold-resistant steels with the Arc index, which belong to strength categories of 315–750 MPa and meet the modern requirements of the Rules of the Russian Maritime Register of Shipping (RMRS–2017) and GOST R 52927–2015 (State Standard of Russia), i.e., the main national standards for the supply of shipbuilding steels, are intended for

Table 6. Cold-resistant steels produced in the largest metallurgical plants for operation in the Arctic region

Strength category, MPa	Cold-resistance category	Maximum rolled thickness, mm	Initial workpiece	Production technology
315	D, E, F	70	Slab	TMP + RC
355	D, E, F	100	Slab	TMP + RC
	Arc	50	Ingot	TMP + RC
390	D, E, F	100	Slab	TMP + RC
	Arc	50	Ingot	TMP + RC
420	D, E, F	60	Slab	TMP + RC
	D, E, F	61–100	Slab	QHR + T
	Arc	50	Slab	TMP + RC
460	D, E, F	60	Slab	TMP + RC
	D, E, F	61–100	Slab	QHR + T
	Arc	50	Slab	TMP + RC
500	D, E, F, Arc	60	Slab, ingot	QHR + T
620	D, E, F, Arc	60	Slab, ingot	QHR + T
690	D, E, F, Arc	50	Slab, ingot	QHR + T
750	D, E, Arc	40	Ingot	QHR + T
890	D, E	50	Slab	QHR + T
960	D, E	50	Slab	QHR + T

use in the design and material supply chain for complex Arctic marine equipment produced at PAO Vyborgskii Sudostroitel'nyi Zavod, AO Baltiiskii Zavod, PAO Severnaya Verf', AO Admiralteiskie Verfi, AO PO Sevmash, and AO Dal'nevostochnyi Zavod Zvezda.

CONCLUSIONS

The development of precision regimes of thermomechanical treatment with accelerated cooling and quenching from the state of rolling heating based on successive grinding of the grain and subgrain structures via the controlled processes of the formation of the main structure (grain growth, recrystallization, fragmentation, polygonization, and phase transitions) is a new stage in the development of the production of highly reliable, cold-resistant steels, especially those with the Arc index.

At the same time, technology has been developed to reduce structural anisotropy, to obtain a given ratio and morphology of structural components, to meet certain requirements for fine structural parameters, and to ensure an acceptable degree of heterogeneity of the structure across the thickness of sheets, which guarantees the high performance of sheet products of various strength level at low temperatures.

Continuous improvement of the technological processes for the production of cold-resistant steels based on fundamental and applied research helps to increase the reliability and safety of the operation of complex marine equipment for Arctic applications.

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